DENDRITIC STEEL

By H. G. Keshian

Abstract

In this paper the author discusses the occurrence, origin, chemical and crystalline nature of dendritic structure and its mechanical properties. He gives the results of a series of tests showing the physical properties, directional strength, wear and fatigue resisting characteristics of a number of dendritic and nondendritic carbon and alloy steels; (1) in as cast condition, (2) after normalizing, (3) after hot working and heat treatment.

THE writer has handled many tons of different types of steel for metal working tools and machine parts much of which showed dendritic structure upon deep etching; and in view of the controversial attitude generally held toward this type of structure, he has rather critically watched the behavior of these tools and machine parts in service. Some of the conclusions which he was able to arrive at were included in a paper on deep etching presented to this society in 1927[1]. In the present paper the matter of dendritic structure has been studied at a greater length.

Due to the fact that the macroetching of iron and steel as a means of inspection is becoming more and more general and the presence of dendritic structure in steel is frequently revealed by this method, often difficulties of a practical nature come up, whether the method of deep etching is used as a means for inspection or as a

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[1]H. G. Keshian, "Deep Etch Test for Iron and Steel," TRANSACTIONS, American Society for Steel Treating, Vol. 12, 1927, p. 689-736.

means for investigation of failures on account of the existence of considerable difference of opinion on the importance of dendritic structure. Dendritic structure, for instance, for some people, is an inherent characteristic of the metal, therefore, of no particular harm. For others, it is a positive proof of the segregation of chemical elements in the steel and results in inferior physical properties. To still another group, it is a solidification phenomenon common to all metals and although there may exist a slight amount of segregation its effect on the general properties of steel is far from being alarming.

It is, therefore, obvious that such radically different opinions on the nature of dendritic structure, even among men familiar with steel metallurgy, indicate the lack of sufficiently definite knowledge of the causes, the nature and the mechanical properties of this type of structure; but in order to realize the importance of the subject, we must bear in mind the fact that the most significant change which the metal undergoes takes place when it passes from the liquid to the solid state, which is the time that the dendrites are formed; because it is here that are contracted, so to speak, the intermolecular alliances among the various elements in the steel which chiefly control its future properties.

GENERAL OCCURRENCE OF DENDRITIC STRUCTURE

The word "dendrite" is derived from the Greek word "dendron" meaning tree-like and is generally applied to a certain type of structure which the metals develop upon solidification. It can be brought out by etching a section of the metal with a suitable solution. In that case it may be visible either to the unaided eye or under a magnifying glass. At times it is visible without etching as shown in Fig. 1.

Dendritic structure, however, is not peculiar to metals alone. It is also found in nonmetallic as well as in metallic substances as shown in Fig. 2. Moreover, dendritic structure does not necessarily develop upon solidification from the liquid state. It may be formed upon solidification from vapor, solution and molten state.

OCCURRENCE OF DENDRITIC STRUCTURE IN STEEL

Dendritic structure can occur in any type of steel, as shown in Fig. 3, but all steels do not develop dendritic structure to the same extent. High carbon steels show more tendency to be dendritic than
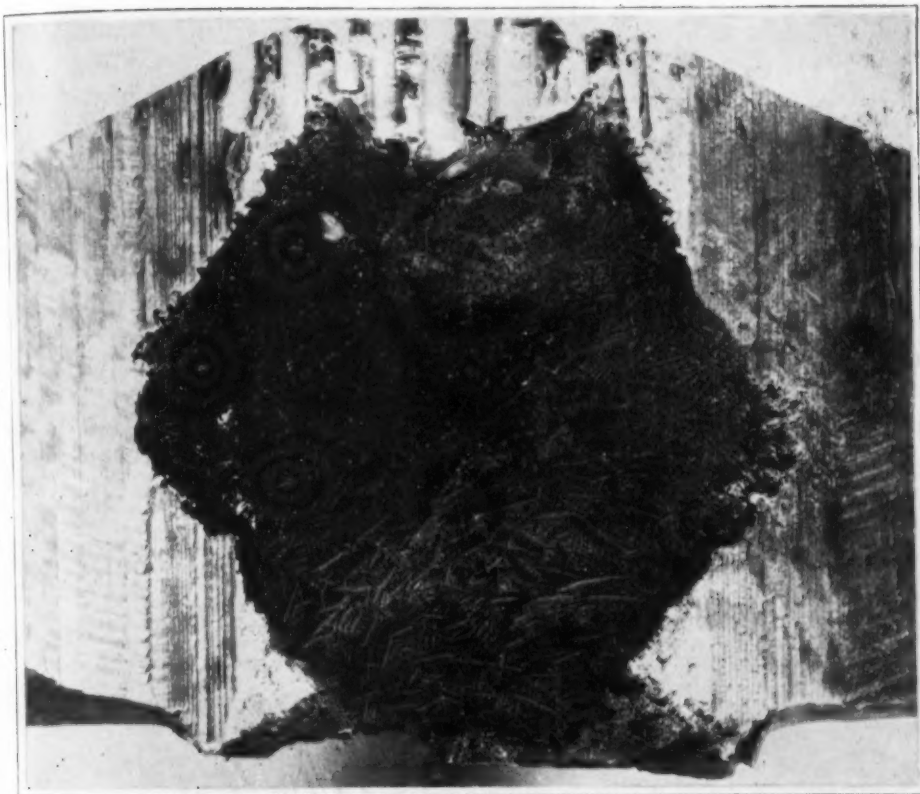
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Fig. 1—Dendritic Structure Formed in the Shrinkage Cavity of Nickel-Chromium Alloy Casting 1⅛ Times Original Size. Nickel, 62 per cent; Chromium, 14 per cent; Carbon, 0.85 per cent.

low carbon steel; nickel and nickel-chromium steels, more than the carbon-chromium and chromium-tungsten steels; etc., but all steels are by no means dendritic and even the same type of steel is not always dendritic.

Occurrence of dendritic structure is independent of the method of melting. Steels made with crucible, electric and open-hearth methods are all liable to develop dendritic structure.

The location of the dendritic area in an ingot and finished bar varies as shown in Fig. 4. It may occur on the outer portion of the ingot or over the entire cross-section of it. At times, two distinct dendritic areas are noted in the cross-section of an ingot, in which both the outside and the center are dendritic with a nondendritic zone between them. This is rather uncommon and is explained under the section of origin of dendrites.

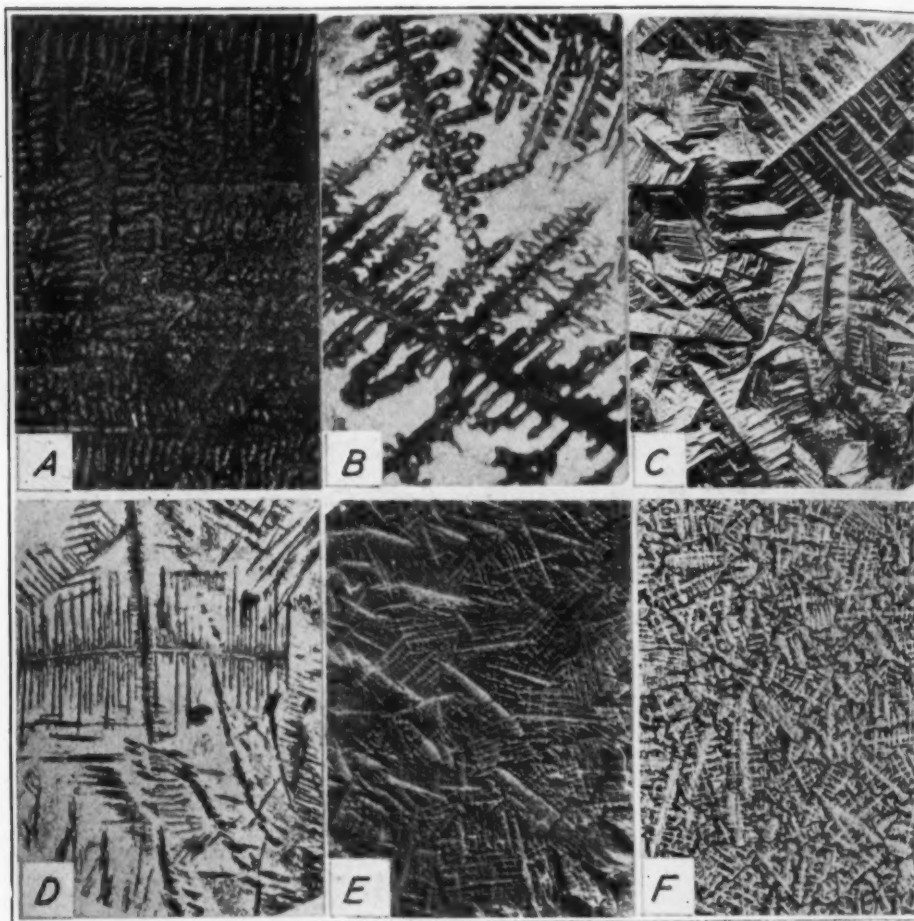The degree of dendritic structure varies considerably from one

Fig. 2—Dendritic Structure in Various Metals and Nonmetals. (A)—Pure Aluminum (Gulliver), (B)—Sal Ammoniac (Rosenhain), (C)—Antimony (Campbell), (D)—Augite × 90 (Rosenbusch), (E)—62 per cent Nickel, 14 per cent Chromium, 0.85 per cent Carbon Balance Iron, (F)—Steel. × 2.

end of an ingot to the other due primarily to the difference in the rate of cooling of the two ends, the dendrites gaining in size toward the hotter end which cools at a lower rate. See Fig. 5.

In the case of a finished bar, the arrangement of the dendritic zone undergoes an apparent change due to hot work. The cross-section of a bar made from an ingot with a narrow zone of dendrites on the outside, after sufficient hot work, appears free from dendrites. If the ingot were wholly dendritic, the bar would show a nondendritic outside surrounding a dendritic interior.

The occurrence of the dendritic structure even in the same type of steel is quite independent of the size of the finished bar, although
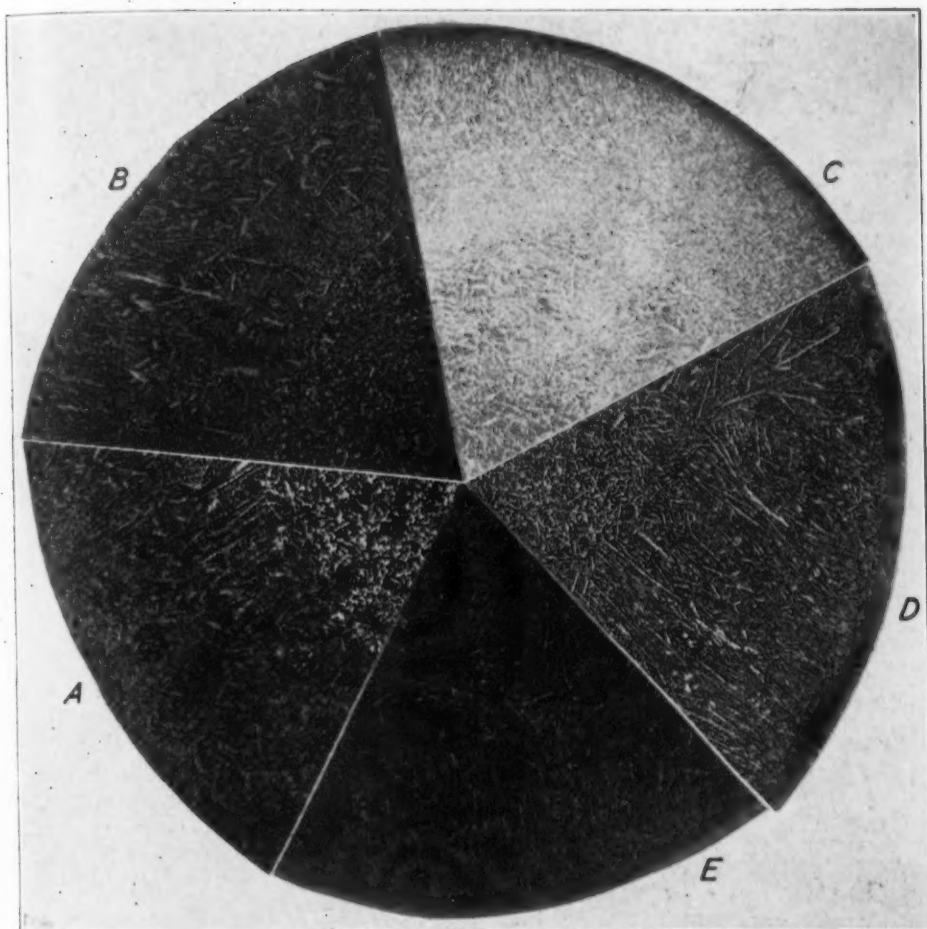
Fig. 3—Dendritic Structure in Different Types of Steels. (A)—Nickel Steel, Nickel, 3 per cent; Carbon, 0.12 per cent; (B)—Carbon Steel, Carbon, 0.60 per cent; Manganese, 0.69 per cent; (C)—Tungsten Steel, Tungsten, 2.58 per cent; Carbon, 1.28 per cent; (D) —Tool Steel, Carbon, 1.07 per cent; Manganese, 0.42 per cent; (E)—Tap Steel, Tungsten, 1.18 per cent; Carbon, 1.18 per cent.

under fairly uniform mill practice the occurrence of dendritic struc-
ture should bear a relation to the size of both the ingot and the fin-
ished bar.

ORIGIN OF DENDRITES

Dendritic structure is formed during the period of solidification,
but so far, no one has been able to directly observe it in the actual
process of formation. Consequently, our conception of how the
dendrites make their appearance in the steel has to be derived, (1)
from a study of the dendrites after the metal has solidified; (2) by
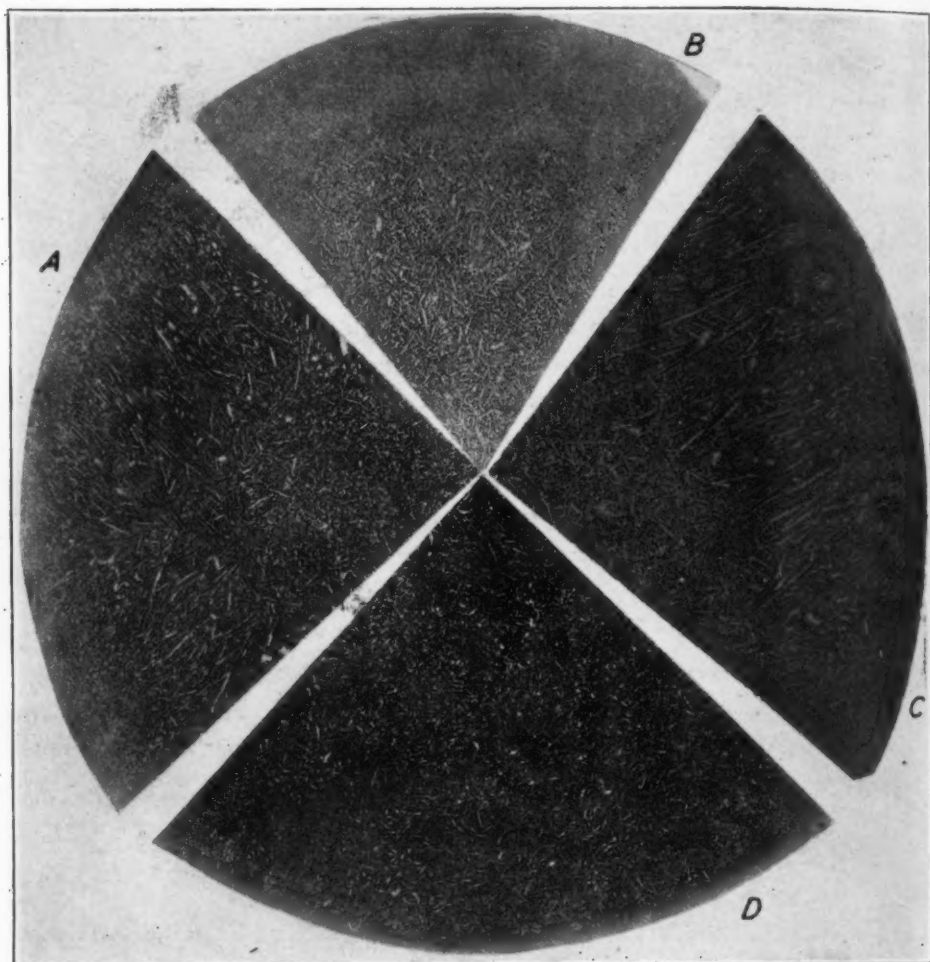studying the mode of dendritic formation of other substances, which

Fig. 4—Types of Dendritic Areas in the Cross-Section of Bar Stock. (A)—Dendritic on the Outside. (B)—Dendritic Area in the Center. (C)—Whole Section is Dendritic. (D)—Dendritic Area on the Outside Separated by a Nondendritic Area from Dendritic Center. (A), (C), and (D) Sections of 6-Inch Bars. (B) Section of 3-Inch Bar. All 1 Per Cent Carbon Tool Steels.

during their solidification are more accessible to such study than steel.

It is generally believed that the dendritic structure in steel is formed as follows:

When the molten steel is allowed to cool, solidification starts at numerous points called "nuclei" or the "centers of crystallization" and advances with great rapidity and in definite directions. As the iron and steel, like most metals, crystallize in the cubic system, the growth takes place in a manner so as to build up a cube. The direction of growth, from the starting points, corresponds to the three principal axes of the cube. In other words, six mutually perpendicu-
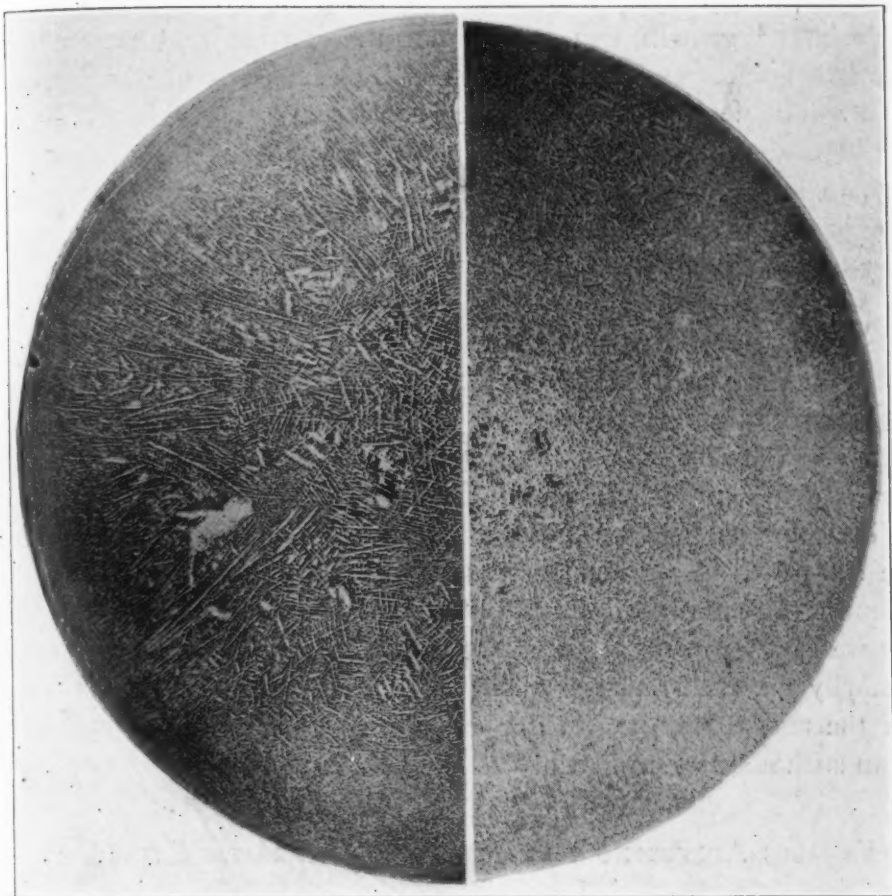
Fig. 5—Photograph of the Opposite Ends of a Bar of 4-Inch Diameter, 1.00 per cent Carbon Tool Steel 11 Feet Long Showing the Variation of Dendritic Structure from One End of the Bar to the Other. Etched in 500 Cubic Centimeters Hydrochloric Acid, 70 Cubic Centimeters Sulphuric Acid and 200 Cubic Centimeters Water. All Macrostructures Given in this Paper are Etched in this Solution Unless Otherwise Stated.

lar solid branches of metal shoot out from each center after the fashion of a 6-armed cross. These six arms are called the primary axes. On these primary axes and at various points, other sets of branches shoot out which are called secondary axes. Branches which grow on the secondary axes are called ternary axes. In the meantime similar activity of solidification goes on from other "centers of crystallization" until they meet each other; so that when the metal is completely solidified an interlocking tree-like structure is the result.

It is obvious that the location of the dendrites will depend on the location of the centers of crystallization. In the case of molten steel cooling in a mold, the centers will naturally first appear on the outer

edge of the solidifying metal where it comes in contact with the cold mold. The growth will start from the outside and proceed toward the center and the dendrites will assume a columnar arrangement. The depth of the dendritic zone formed on the outer edge of the ingot depends on the kind of steel, size of the ingot, etc. Generally, in smaller ingots, the dendritic zone is small, varying from a fraction of an inch to several inches, and it may reach the center in very large and slowly cooled ingots.

The occurrence of a dendritic area in the center, separated by a nondendritic zone from the dendritic area on the outside at times is noted on large bars. This can be explained by assuming that some of the dendrites in very large ingots may break away from the dendritic zone and fall into the bottom of the still liquid metal. This assumption which is used by Howe to explain why the lower central axis of large ingots is purer than the average, will not hold true here if the central dendritic zone occurs all along the length of the ingot. Nor the presence of another dendritic zone immediately surrounding the wall of the pipe, as pointed out by Giol ri,[2] could explain satisfactorily the occurrence of a dendritic zone in the center of the bar for the reason that no possible existence of pipe could be suspected from such a structure that has come under the writer's observation.

FACTORS AFFECTING THE DEGREE OF DENDRITIC STRUCTURE

The degree of dendritic structure is governed by the number and the size of dendrites, which seem to depend principally on, (1) casting temperature of the steel; (2) chemical composition; (3) rate of solidification.

1. Casting Temperature—If we cool a mass of molten metal suddenly, a center of crystallization will immediately start at almost every point in the metal; but as the starting and stopping of solidification would occur almost instantaneously, the growth at each point will also stop instantaneously. This will result in a structure consisting of exceedingly small, arrested dendrites which cannot be seen without a microscope. Such a structure obtained by cooling molten high speed steel in water is shown in Fig. 6[3]. This condition of cool-

[2]F. Giolitti, "Heat Treatment of Soft and Medium Steels," McGraw-Hill Book Company, 1921.

[3]F. Rapatz, "Influence of the Structure As Cast Upon Manufacturing and Qualities of Some Alloyed, Especially High Speed Steel," TRANSACTIONS, American Society for Steel Treating, Vol. 13, 1928, p. 1009-1022.
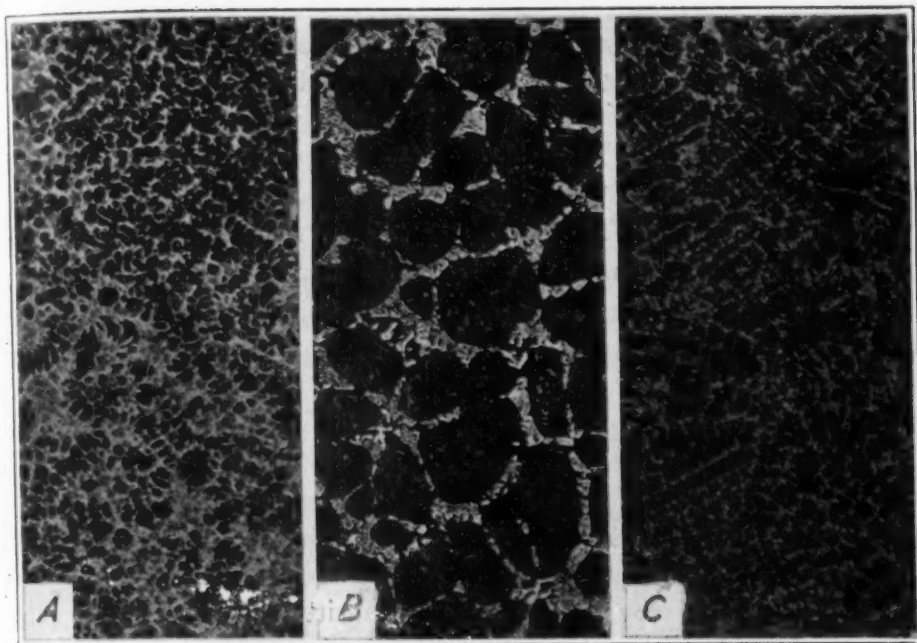
Fig. 6—Photomicrograph (A) is High Speed Steel Quenched in Cold Water from the Molten State. × 100. Photomicrograph (B) is Network in the Center of an Ingot of High Speed Steel. × 100. Photomicrograph (C) is the Network in an Ingot of High Speed Steel Near the Surface. × 100. (Rapatz)

ing, however, does not obtain in practice and a molten mass of steel always commences to solidify in some parts earlier than others and in the case of a steel solidifying in a mold, much time elapses between the starting and stopping of solidification which allows some time for the dendrites to grow. Obviously, the length of time that elapses between the starting and stopping of the solidification, in a given size of ingot, will depend on the amount of heat that is to be extracted from the interior of the mass; but as the amount of heat of a given mass of metal depends on its temperature, the higher the casting temperature the longer the time to solidify and the larger the dendrites.

(2) Effect of the Chemical Composition of the Steel on the Development of Dendrites—The chemical composition of the steel undoubtedly has an effect on the development of dendritic structure; because we know that some steels show more tendency to dendritic growth than others; but as there is not sufficient experimental data yet showing the specific effect of various elements, any comments that can be made now will be mostly of a speculative nature.

Chemical composition seems to affect the intensity of dendritic

structure, in the following ways, (a) by increasing or decreasing the centers of crystallization, (b) by affecting the viscosity of the molten metal; (c) by affecting the surface tension of the molten metal; and by other ways.

(a)   It is generally accepted that the presence of foreign elements in a metal affects its grain size both during the crystallization and recrystallization.   This phenomenon can reasonably be explained by a change in the number of centers or the nuclei of crystallization. The effect of nickel in reducing the grain size of cast iron and the effect of small amounts of iron on reducing the grain size of alpha brass and retarding the rate of its crystallization are common examples.   The absence of dendritic structure in steels badly contaminated with impurities is another example.

(b)   The part played by various elements in the development of dendrites by affecting the viscosity of the molten steel is little known; but it is evident that after the solidification has started at the various centers of crystallization the dendritic branches grow with great rapidity by the quick addition of new crystals.   The size of dendrites built up in this manner will essentially depend on the unobstructed action of the atoms and the velocity with which they can move.   This will depend on the fluidity or the viscosity of the molten metal.   The viscosity of molten metals is generally very low and at an average, perhaps, not more than one fifth of the viscosity of boiling water, which, perhaps explains to a great extent the rapidity with which the dendrites are formed.   They are so rapidly formed, that we cannot prevent their formation, however small, even by quenching the molten steel in water.[3,4]   This low viscosity of metals would perhaps also explain the fact that almost all metals are crystalline. While the minerals have high viscosity and the crystalline structure among them is not as general as in the metals, and the amorphous structure which is absent in metals, is frequently found in minerals. If the viscosity of the molten rock is exceedingly high the mineral solidifies in amorphous form, free from any crystals.   For instance, a molten rock or magma of certain composition upon cooling solidifies as the mineral obsidian, being glassy and free from crystalline structure; but when a small amount of water and fluorine is present in the magma they materially lower its viscosity and thus promote selective

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[4]N. T. Belaiew, Written discussion of Dr. Sauveur's paper, "Crystallization of Iron and its Alloys," TRANSACTIONS, American Society for Steel Treating, Vol. 5, 1924, p. 219-221.

crystallization, then the magma crystallizes as granite. Such con-
spicuous examples are absent in the case of steel, but it would appear
that the elements which influence the viscosity of the molten steel
should also affect the development of dendrites. By increasing the
viscosity of the molten steel, if that is possible, we should be able to
reduce the dendritic growth.

The effect of specific elements on the viscosity of steel is not
sufficiently clear other than the general assumption that the greater
the amount of elements in the steel the lower the melting point of
the alloy, therefore, the lower the viscosity. The presence of iron ox-
ide is believed to lower the viscosity and in those steels higher in
oxygen content it is believed that they show greater dendritic struc-
ture than those steels low in oxygen. The effect of silicon is to in-
crease the viscosity of the steel and the equiaxed character of the
crystals in silicon steels is perhaps due to its increased viscosity.

(c) The influence of surface tension on the solidification of
substances has now and then been disputed, but the convincing evi-
dences put forward by the pioneer work of Quincke still is one of the
outstanding theories of the crystallization of metals and according
to Dr. Desch the surface tension plays an important part in the solidi-
fication of metals.[5] It is relatively simple to follow the effect of
alloying two pure metals of greatly varying surface tensions on the
crystalline habit of the alloy, but the possible effect of surface tension
on the development of dendrites in steel becomes altogether too com-
plex and too theoretical to discuss here.

(3) *Rate of Solidification*—In the development of dendritic
structure the rate of cooling is perhaps the greatest factor. If the
steel is allowed to solidify very slowly, comparatively few centers of
crystallization will be set up and the resulting dendrites will be of
larger size; but if the cooling is rapid the number of centers of
crystallization will be greater, the period of solidification will be
shorter and a greater number of small dendrites will be formed. From
this, one may assume that with sufficiently rapid cooling the formation
of dendrites can be totally prevented. This, although a familiar
phenomenon in minerals, has never been noted in metals. Bekier
cooled antimony to —70 degrees but it still showed centers of crystal-
lization.[6] N. T. Belaiew cooled molten steel in water and still found

[5]C. H. Desch, "Solidification of Metals from the Liquid State," *Journal*, Institute of
Metals, 1914, N. 1, p. 57-118.

[6]Ibid. p. 92.
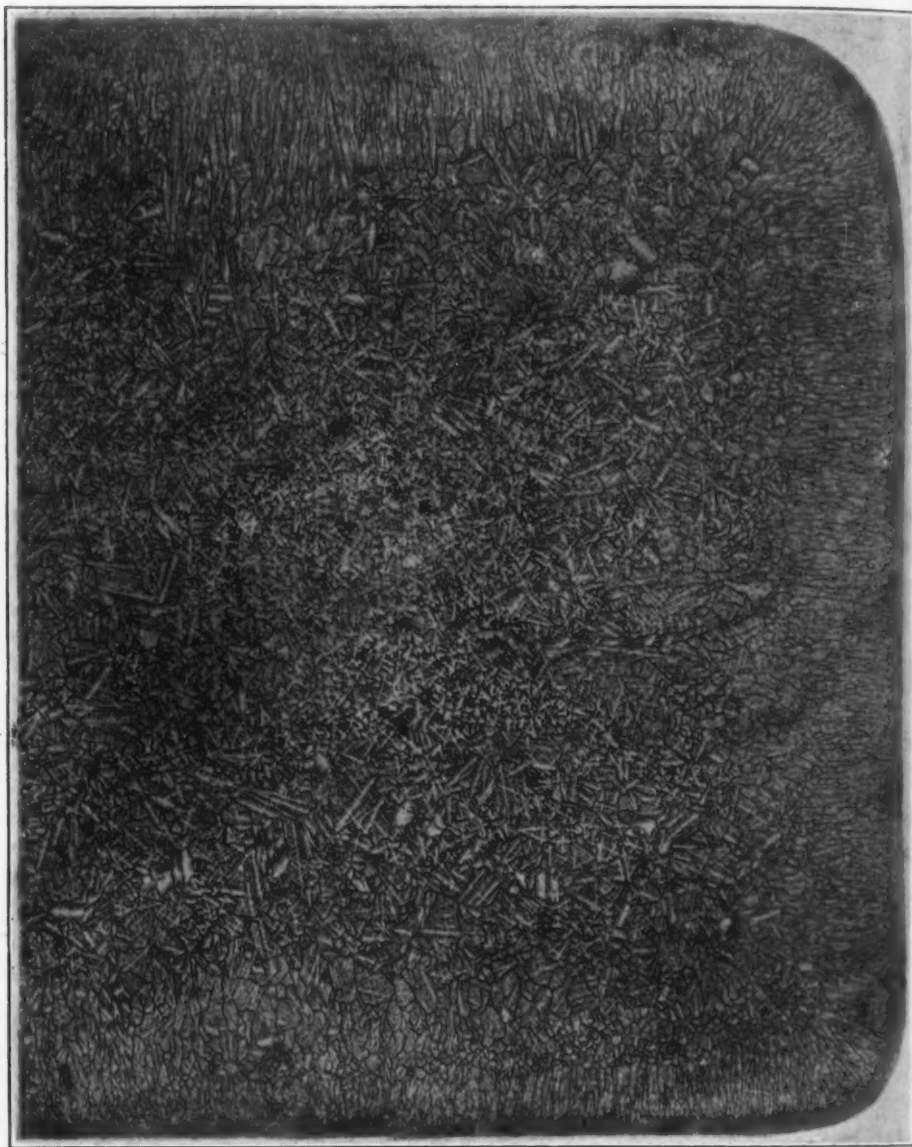
Fig. 7—Cross-section of 6 by 6-Inch Straight Carbon Steel Ingot Cooled Slowly in the Mold which was Buried in Lime. It Shows the Effect of the Rate of Cooling on the Development of Dendritic Structure. Same Analysis as the Ingot Shown in Fig. 8— Carbon, 1.05 per cent; Manganese, 0.24 per cent; Silicon, 0.20 per cent; Phosphorus, 0.027 per cent and Sulphur, 0.009 per cent.

perfectly formed dendrites.[7] In Fig. 6 is shown the structure of tungsten steel obtained by cooling the molten metal in water. In Figs. 7 to 10 is shown the effect of the rate of cooling on the development of dendritic structure in high carbon tool and chromium-nickel steels
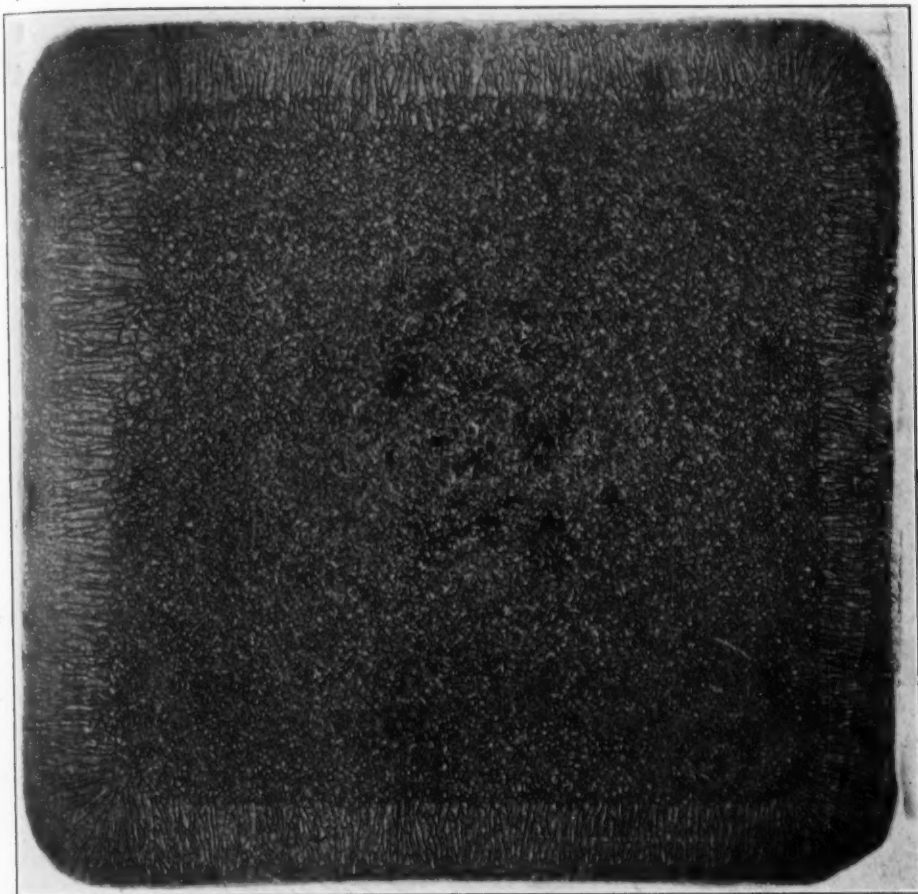
[7]Ibid. p. 297.

Fig. 8—Cross-section of 4 by 4-Inch Straight Carbon Steel Ingot Rapidly Cooled by Setting the Mold in Water. It Shows the Effect of the Rate of Cooling on the Development of Nondendritic Structure. It has same Analysis as the Ingot Shown in Fig. 7.

which were obtained by casting them in 4-inch molds set in water and in 6-inch molds slowly cooled by burying them in lime. From Figs. 6 to 10 it is evident that the degree and the zone of dendritic structure can be greatly reduced by increasing the rate of cooling; but the formation of dendrites of microscopic size can never be prevented, at least, with the methods with which we are familiar now.

Although with a minimum casting temperature and reasonably rapid rate of cooling the formation of dendrites can be materially reduced, we still have the problem presented by the size of the ingot. The greater the ingot diameter, the longer will be the time that will elapse between the starting and stopping of crystallization. Therefore, the larger would be the size of the dendrites and deeper the zone of dendritic area from the outside toward the center. This would

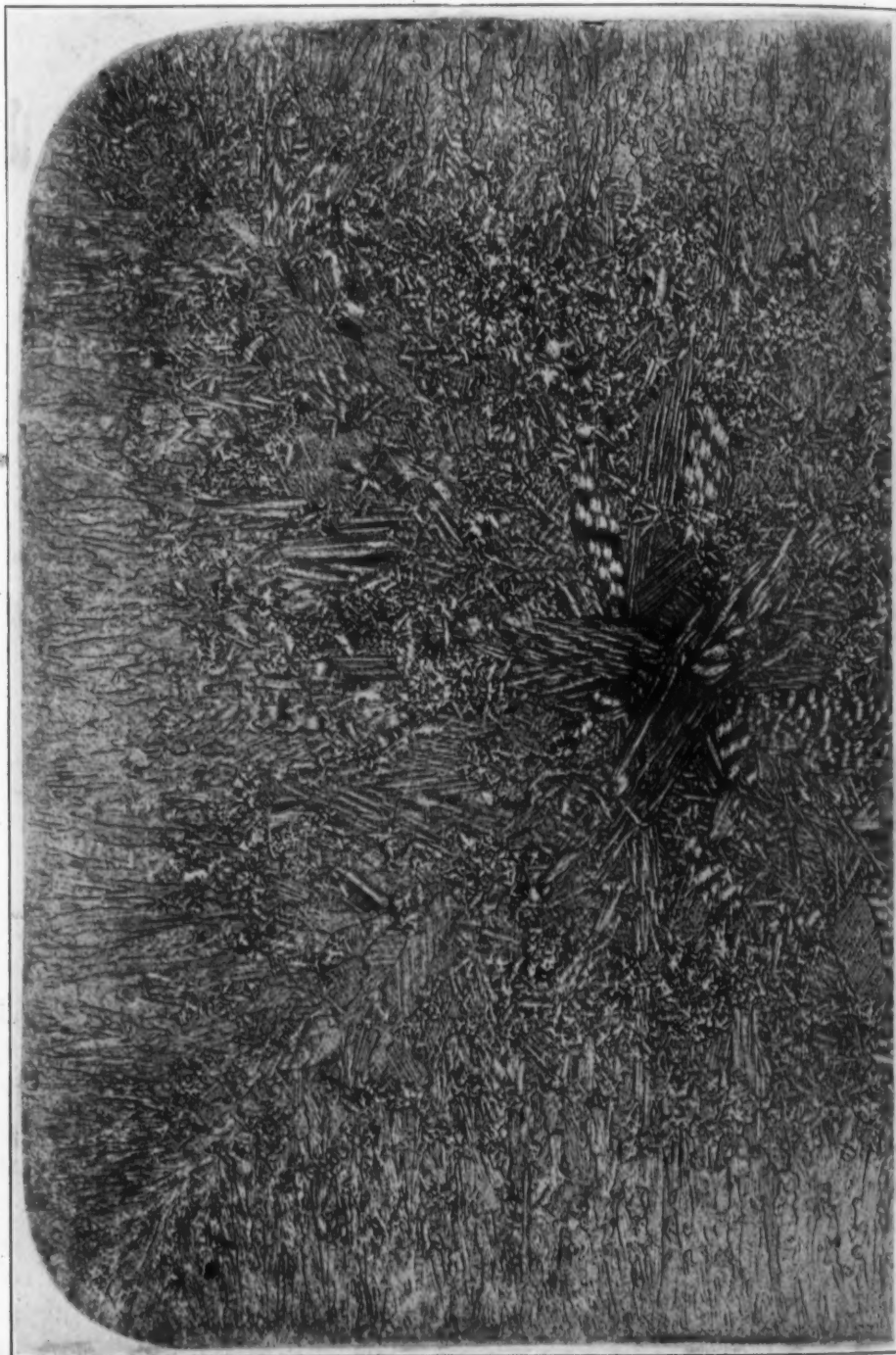Fig. 9—Cross-section of 6 by 6-Inch Low Carbon, Chromium-Nickel Steel Ingot, S. A.E. 3312 Steel, Slowly Cooled by Burying the Mold in Lime. It Shows the Effect of the Rate of Cooling on the Development of Dendritic Structure. Compare with Fig. 10. Analysis: Carbon, 0.125 per cent; Chromium, 1.43 per cent; Nickel, 4.19 per cent; Silicon, 0.23 per cent; Phosphorus, 0.025 per cent and Sulphur, 0.010 per cent.
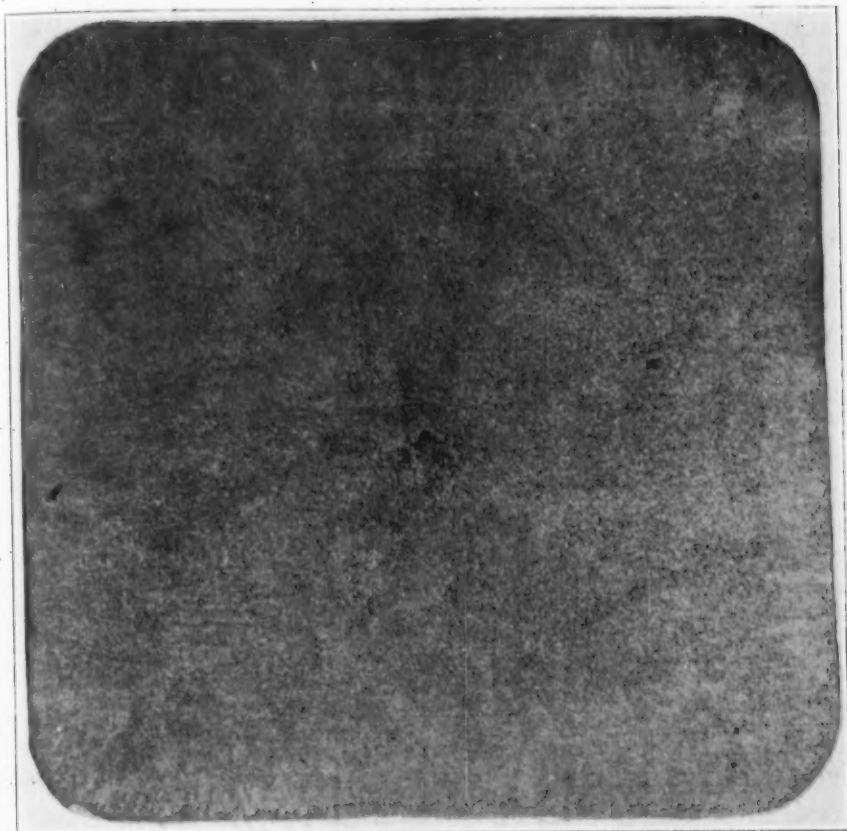
Fig. 10—Cross-section of 4 by 4-Inch Low Carbon, Chromium-Nickel Steel Ingot, S.A.E. 3312 Steel Rapidly Cooled by Setting the Mold in Water. It Shows the Effect of the Rate of Cooling on the Suppression of Dendritic Structure. Compare with the Ingot of Same Analysis Shown in Fig. 9.

mean that when a minimum degree of dendritic structure is desired the lowest possible casting temperature, a rapid rate of cooling and a minimum size of ingot must be used.

CRYSTALLINE NATURE OF DENDRITES

The structure of iron and steel like that of other metals is made up of three principal units, namely, atoms, crystals and grains. In order to better understand the crystalline structure of dendrites, let us briefly consider the relation of these units to each other and to the general structure of the metal.

In their simplest conception, atoms make up the crystals, crystals make up the grains and the grains make up the structure of the metal in the following manner. In a molten mass of metal, the atoms are in a dispersed state; but when the former reaches the freezing point,

Fig. 11—Dendritic Grains or "Patches" in Slowly Cooled 6 by 6-Inch Carbon Tool Steel Ingot. Same as in Fig. 7, Magnified Twice the Original Size.
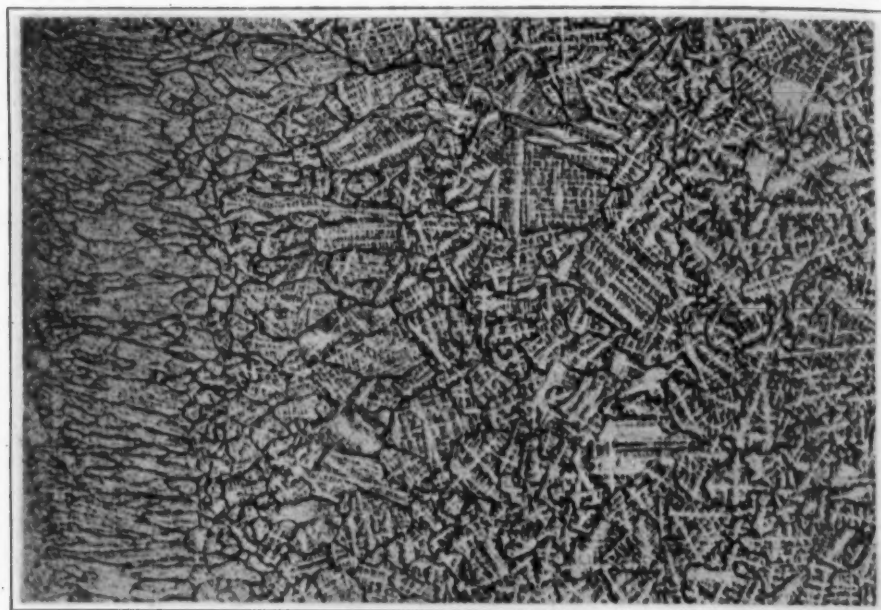
a group of "pioneer atoms" at numerous centers, get together and take up definite positions and build up the first crystals. This is followed by fresh groups of atoms which build up more crystals around the first ones. After the formation of the first crystals, the building up of the additional crystals takes place with such rapidity that a competitive and feverish building activity, so to speak, goes on incessantly from each center of solidification until the whole mass of the metal has been solidified. When the solidification of the metal is complete, the resulting structure, when viewed on a section, appears as if made up of areas or patches of varying patterns, see Fig. 11. These patches or the "crystal communities" are the dendrites which in reality are the third structural unit of a cast metal, namely, the grains.

Now, let us consider the crystalline nature of these three structural units and the changes which they undergo after the solidification: but for a clearer understanding, perhaps somewhat of an elementary treatment of this phase of the subject would not be superfluous.

The first set of atoms which get together at the various centers of solidification and build up the first crystals, are not actually in contact with each other; but separated by very minute and regular distances. The positions which the atoms take up in the space in respect to one

another, determines the shape of the crystals. In some metals the atoms arrange themselves in such a manner as to lay out the form of a cube; in others, octahedrons, hexagons, etc. When the atoms arrange themselves in the cubical form, the crystals which they thus build up, may be a body-centered or face-centered cube. In the body-centered cube there is one atom at each corner of the cube and one in the center of the cube, making altogether nine atoms. In the face-centered cube, there is one atom at each corner of the cube and one atom at the center of each of six faces of the cube; but no atom at the center of the cube, making fourteen atoms in all. These atoms are held in space by forces of attraction and repulsion and any attempt to change their position, such as to deform the metal, meets the resistance of these forces.

Iron and steel crystallize in cubic and octahedral systems, in which, according to Howe, the cubes build up the octahedrons. Therefore, the first atoms which build up the first crystals at each center of solidification arrange themselves as to form a cube and the succeeding crystals arrange themselves after the position or the orientation of the first crystals: but as all of the "first crystals" do not hold the identical positions toward each other, the resulting colonies of crystals which build up the dendrites, will have different orientation and would appear like different patches as in Figs. 11 and 12. From Fig. 12 it will be seen that it is the individual position which the "first crystals" take up in respect to each other, that gives the different appearance to the grains. It must be remembered that the growth of all the grains does not take place with the same speed. Neither the growth from each center progresses in all directions in the same rate in the same grain. For that reason the resulting grains or the dendrites become of varying sizes and outlines.

When the atoms are building up the crystals and the crystals are building up the grains as the atoms give the geometrical outline to the crystal, the crystals in turn strive to give a geometrical form to the grains by faithfully following the crystal system to which the metal belongs. In the case of iron and steel the crystals when building up the dendritic grains tend to build up an octahedral grain by laying out the three principal axes as stated elsewhere; but as the conditions for the formation of perfect crystals do not exist in the case of a solidifying mass of steel in a mold, the crystals cannot carry out their plan to build up perfect cubes or octahedrons and the dendritic grains instead of having the shape of perfect cubes or octahe-
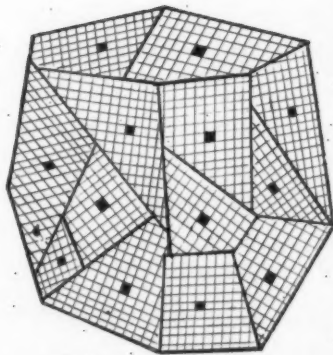
Fig. 12—(A) Schematic View of the Formation of Dendrites as Caused by the Difference in the Orientations of the "First Crystals" (Dark Squares) which Build up the Dendritic Grains by Giving their Own Orientation to the Crystals that Follow (White Squares) them. (B) Dendritic Grains as Developed by Slow Cooling Low Carbon Nickel-Chrom.um Steel as Shown in Fig. 9 and Illustrates how Dendritic Grains Result from a Difference in the Orientation of the "First Crystals" as Depicted in Fig. 12A. × 2.
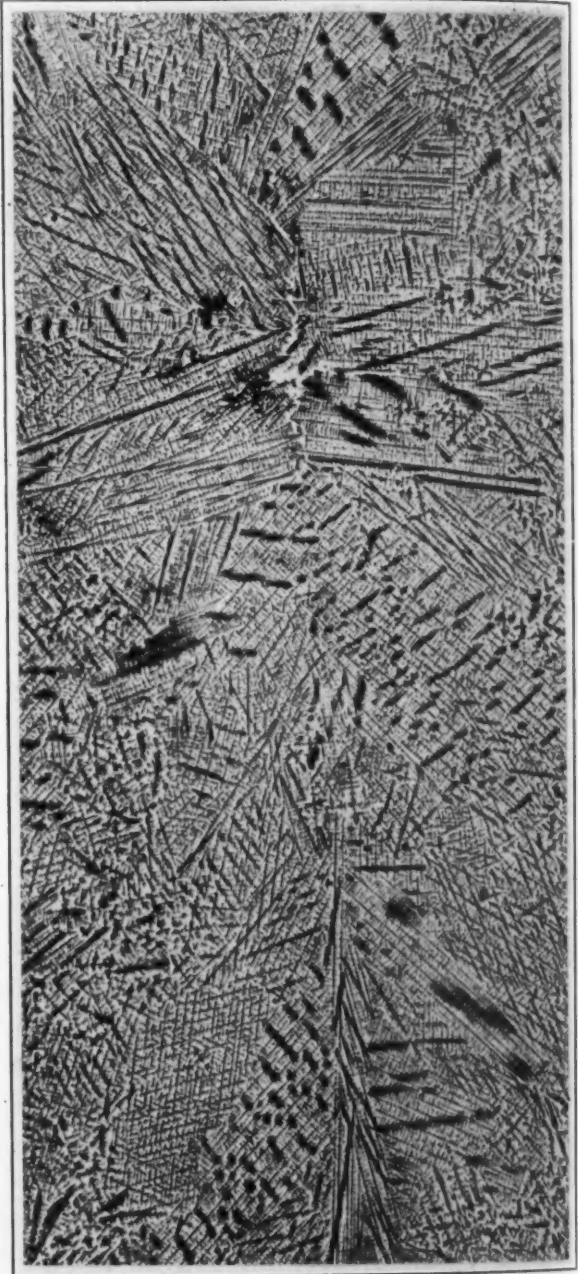
drons, represent the skeleton of these crystals. Dendrites, therefore, are the skeletons of cubes or octahedrons, built up by the minute crystals in which the arrangement of the orientation of crystal colonies vary from one dendrite to another.

STRUCTURAL CHANGES IN DENDRITES

The crystalline structure which was described in the preceding paragraphs is the structure which the dendrites acquire immediately upon solidification; but as the steel cools down to room temperature certain changes appear to take place in the structure of dendrites which we will consider now.

Structurally, the first crystals which are formed at each center of crystallization and the crystals that follow them are body-centered cubes. In pure iron and in steel containing about 0.36 per cent carbon, a body-centered lattice is formed and that lattice is retained down to a temperature of about 2640 degrees Fahr. (1450 degrees Cent.). Then the structure changes to face-centered cubes until a temperature of 1650 degrees Fahr. (900 degrees Cent.) is reached. Below 1650 degrees Fahr. it changes back to body-centered cubic lattice. These three stages correspond to the delta, gamma and alpha phases.

Following these lattice changes any structural change which takes place is of atomic magnitude and confined to the crystals which make up the dendritic grains and therefore have almost no effect on the general outline of the dendrites; because if these atomic movements alter the common orientation of the crystals in one dendritic grain, they also alter the common orientation of the crystals in other dendritic grains which leaves the original dissimilarity among various dendrites the same. This will lead us to another conclusion that any structural changes which one might expect to bring about in the dendrites by changing the lattice arrangement of their crystal components, by means of heat treatment, that change will be confined to the crystals composing the dendrites, and will not have any affect on the general outline of the dendritic grains. This is just what we find in practice and we cannot eliminate the dendrites by heat treatment.

These changes which only effect the inner structure of dendritic grains, are, however, quite important and have been thoroughly discussed by N. T. Belaiew in his various papers.[8] According to Belaiew, as the steel solidifies from the molten state to room temperature it passes through three stages—the zone of primary crystallization, the zone of granulation and the zone of secondary crystallization. These three zones are shown in Fig. 13. In the zone of primary crystallization the dendrites are formed and the solidification is complete.

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[8] N. T. Belaiew, "The Granulation Hypothesis and the Delta-Gamma Change in Iron, Carbon and Iron-Nickel Alloys," TRANSACTIONS, American Society for Steel Treating, Vol. 5, 1924, p. 549-570.
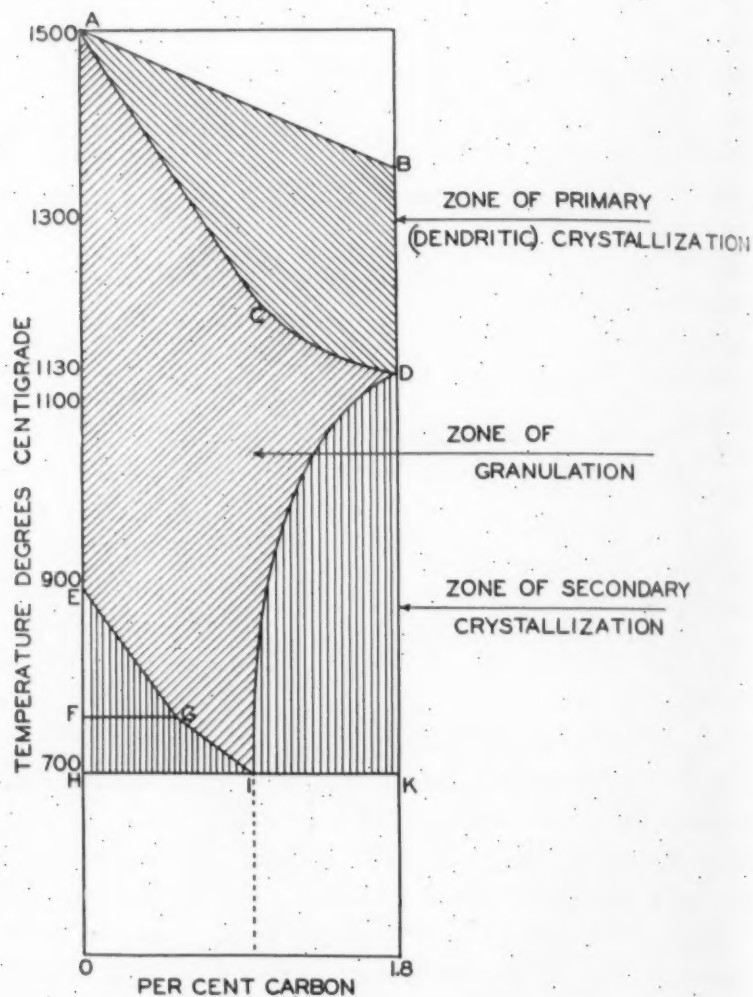
Fig. 13—Part of Iron-Carbon Diagram Showing the Three Zones of Structural Changes Occurring in Dendrites According to Belaiew, (Sauveur).

In the zone of granulation, the dendrites according to Belaiew, break down to small grains, leaving the outer form of the dendrites unaltered. In the zone of secondary crystallization the excess ferrite in hypoeutectoid and excess cementite in hypereutectoid steels separate out and arrange themselves in three distinctive manners, which Belaiew calls, (1) structure of large crystals, in which the excess element lodges itself along the dendritic axes. This occurs by rapid cooling through the zones of granulation and secondary crystallization. The characteristic of this structure is that the excess element shows up as discontinuous patches. (2) Widmanstatten structure in which the excess elements, ferrite or cementite, appear in criss-cross outlines.

This structure occurs when the steel cools slowly through the zone of granulation and rapidly through the zone of secondary crystallization. Excess ferrite or cementite by failing to reach the boundaries of the grains, lodge themselves along the cleavage planes of the octahedral grains. The characteristic appearance of this structure is the triangular patterns formed by the intersecting plates of the excess elements. (3) Network structure in which the excess ferrite or cementite appears in continuous meshes. This structure occurs by fast cooling through the zone of granulation and slow cooling through the zone of secondary crystallization in which the excess elements are given more time in the zone of secondary crystallization to reach the boundaries of the grains. The characteristic of this structure is the continuity of the excess elements.

It is to be noted, therefore, that, according to Belaiew, after the dendrites have been formed two fundamental changes take place in their inner structure—granulation and secondary crystallization which we will consider now with a little more detail.

GRANULATION

As stated before the structure of dendrites after they have been formed break up into a number of smaller grains. In other words the microstructure of a dendrite as we find it under the microscope is not the same structure which the dendrites had at the time they were formed. Therefore the inner structure of dendrites must have undergone a structural change subsequent to their formation.

Now, what proof do we have that this is the fact? In the first place we must know the original inner structure of the dendrite at the time it was formed and compare it with the inner structure of the dendrite as we find it under the microscope; but as we do not know directly the original structure of the dendrite we must base our knowledge on our conception of how the dendrites were formed. Our conception of the formation of dendrites as depicted in Fig. 12 is that the first atoms get together, so to speak, and build up the first crystal and the additional crystals in an orderly manner and with similar orientation build up the dendritic grain. This gives birth to a crystalline grain of definite internal structure and made up of small units of the same orientation.

Now, when we actually examine the internal structure of a dendrite under the microscope we find that the structure instead of being
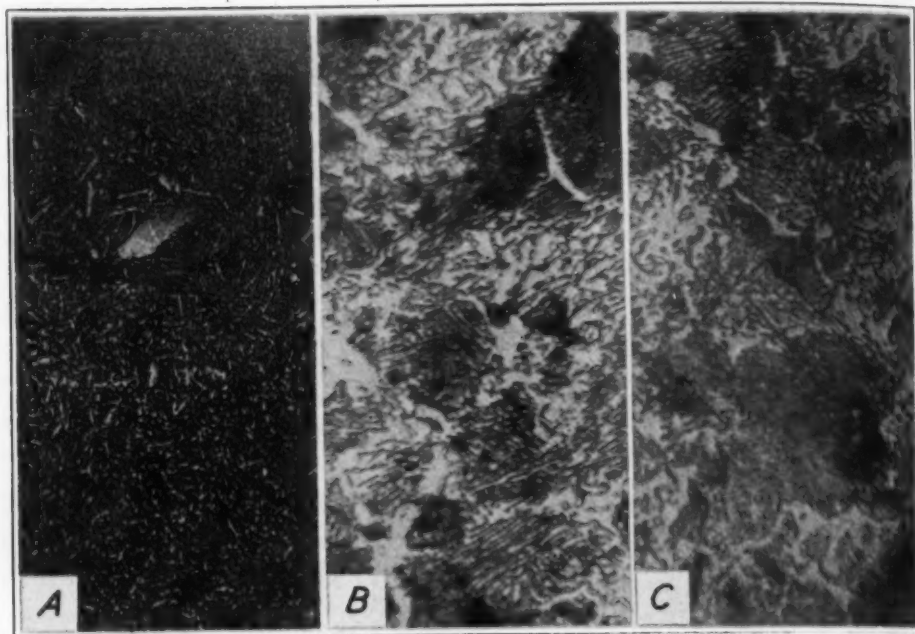
Fig. 14—(A) Macrostructure of an 8-Inch Diameter Bar Stock Containing 0.60 per cent Carbon and 0.69 per cent Manganese Showing one of the Large Dendrites.   Full Size (B) Microstructure of the Large Dendrite Shown in (A) × 400.   (C) Microstructure of Groundmass of (A). × 400.

made up of the minute crystalline units of common orientation is made up of independent and small groups of different orientation. We are, therefore, forced to conclude that this grouping and rearrangement of the inner structure of the dendrites must have taken place after the formation of the dendrites.

That this seems to be the fact will be seen by referring to Fig. 14 where it will be noted that the large single dendrite in Fig. 14A, which must have been a single crystalline unit made up of many tiny crystals of similar orientation, is now divided into many pearlitic grains of dissimilar orientation as shown in Fig. 14B. This being the case then, in what region of the iron-carbon diagram does the rearrangement of the inner structure of the dendrites take place?

According to Belaiew this change takes place in a region between the upper critical range and the upper portion of the austenitic zone, but the proof which he offers, however, is rather inferential and is essentially based, as the author understands, on the assumption that a crystalline change must follow any lattice change or allotropic transformation.

Even granting this assumption to be true, the lattice change as

we know it now can only partly explain the granulation; because according to Belaiew, the granulation occurs in the upper region of the gamma iron zone and he traces it up to 2050 degrees Fahr. (1120 degrees Cent.) and according to Westgren[9] delta iron exists above 2650 degrees Fahr. (1450 degrees Cent.) and below that it changes to gamma iron. In other words a change takes place from body-centered cubic lattice to face-centered lattice below 2650 degrees Fahr. (1450 degrees Cent.) or the granulation starts below this temperature and the upper critical range. But according to Ruer and Klesper[10] this lattice change in the austenitic zone ceases for steels with carbon over 0.36 per cent. Therefore, as Belaiew himself concedes, the lattice change can account only for the granulation of steels with carbon content 0.36 per cent and less.

SECONDARY CRYSTALLIZATION

The structural changes which take place in the crystalline structure of dendrites between the lower and upper critical ranges, unlike the granulation, are quite well known under the term of secondary crystallization, but the conditions under which the types of structure such as Widmanstatten and network structures form are not sufficiently understood. The conditions which give rise to those types of structure, according to Belaiew, have already been mentioned which, however, are not generally accepted by others. For instance the fundamental assumption for the formation of Widmanstatten and network structure, as conceived by Belaiew is that, excess ferrite or cementite under slow cooling migrate toward and to the grain boundaries. This assumption has been questioned by other investigators on the theoretical ground that the migration of excess elements in a mass of already solidified metal is more than the atomic forces or diffusion are capable of doing.[11] Krivobok in experimenting with 0.40 to 0.50 per cent carbon steel by various combinations of slow and fast cooling through the zones of granulation and secondary crystallization obtained a network structure in all cases and failed to confirm Belaiew's conclusions.

[9]A. Westgren and Gosta Phragmen, "X-Ray Studies on the Crystal Structure of Steel," *Journal*, Iron and Steel Institute, Vol. CV, No. 1, 1922, p. 241 and Vol. CIX, No. 1, 1924, p. 171.

[10]R. Ruer and R. Klesper, "Die Gamme-Delta unwandlung de Reinen Eisens und Ihre Beeinflussing Durch C, Si, Co and Cu," *Ferrum*, Vol. 2, 1914, p. 257-261.

[11]F. Giolitti, "Relationship Between Dendritic Structure and Ferrite Mesh," *Chemical and Metallurgical Engineering*, Vol. 22, 1920, p. 921-928.

Again in Fig. 23D is shown the microstructure of the hypereutec-toid steel ingot shown in Fig. 8 which was cooled rapidly through the zones of granulation and secondary crystallization. According to Belaiew the excess element—the cementite in this case—should have arranged itself in discontinuous patches; but from Fig. 23D it is evident that this is not the case and the excess cementite here appears in continuous network structure. It would seem, therefore, that further investigations must be made to determine just under what conditions besides cooling these different structures may be formed.

But so far as the crystalline nature of the dendrites go the important thing to bear in mind seems to be the fact that the final internal structure of the dendrite is not the original structure which they assume or we conceive it to be at the time of their formation, but it undergoes an important change when the metal cools down to room temperature. Although this does not alter the outward appearance of the dendrite, it imparts a different characteristic to the metal than the one we conceive from the accepted manner of the formation of dendrites. In other words we conclude that when the dendrites are formed the crystals which build up the dendrites all have the same orientation in a given dendrite. Therefore, the individual dendrites have different physical properties in different directions, but after they have gone through the structural changes each individual dendrite breaks up into small units of different orientation and then they acquire almost the same properties in all directions, making the metal of better physical properties than it would be without the internal changes.

CHEMICAL NATURE OF DENDRITES

The majority of the existing opinions on the chemical nature of dendrites is that they are caused by the segregation of the chemical elements in the steel. Among the principal evidences offered are the following:

(a) *Etching Characteristics of Dendrites.*—Upon etching a prepared sample of a dendritic steel, the dendrites are brought out quite clearly by the selective action of the etching reagent. It is concluded that if the structure were homogeneous throughout, the etched surface would also be homogeneous. It is the difference in chemical composition that results in selective etching.

This belief, however, that no selective etching can take place without chemical variation is not true. The difference in the orienta-

tion of crystals or the arrangement of atoms can equally cause selective etching. It is a fact that chemical action takes place more rapidly along the planes of loosely packed atoms than along the planes of closely packed atoms. For instance, in the pure graphite the atoms in the planes parallel to the hexagonal base are 1.42 Angstrom units apart; while along the vertical planes they are 3.42 Angstrom units apart and the corrosion is much more rapid in the direction parallel to the basal planes than in the directions parallel to the vertical planes. A similar effect has also been noted in biotite mica.[12] Furthermore, it is the writer's observation that a fairly well developed dendritic structure in steel can be detected by the unaided eye by merely smooth machining the metal without further polishing or resorting to etching, showing a structural difference and not necessarily a chemical difference. Surely, chemical difference can cause structural difference but structural difference is not always caused by chemical difference. If it did, how would we explain the dendritic structure in pure metals, except by a difference in crystalline arrangement?

(b) Another proof offered is that the steel contains various elements that have different melting points. Therefore, during the solidification the elements of highest melting point solidify first and make up the primary axes of the dendrites followed by material of lower melting point, and so on, until the whole mass has solidified, causing a chemical variation from the center to the outside. This view, of course, is principally true, both from the standpoint of the phase rule and constitutional diagrams for alloys, but a very important fact is not to be overlooked, that the selective crystallization does not follow the order of fusibility, but the order of solubility. Examples of this may be taken from both minerals and alloys. In a rock mineral containing pyroxene and quartz, pyroxene crystallizes first, then quartz: while pyroxene is readily fusible, quartz is infusible, yet the latter crystallizes the last. Likewise iron phosphide has a much lower melting point than the iron, yet it does not solidify last but goes into solid solution with ferrite and solidifies at a temperature much higher than its own. If it did solidify in compliance with its melting point alone it should separate out and not go into solution with ferrite. It is obvious that in the discussion of dendritic formation, if solubility is given the consideration which it deserves the idea of segregation will be less dominant and will be relegated to its proper place.

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[12] C. H. Desch, "Chemical Properties of Crystals," *Metal Industry*, 1928, p. 497.

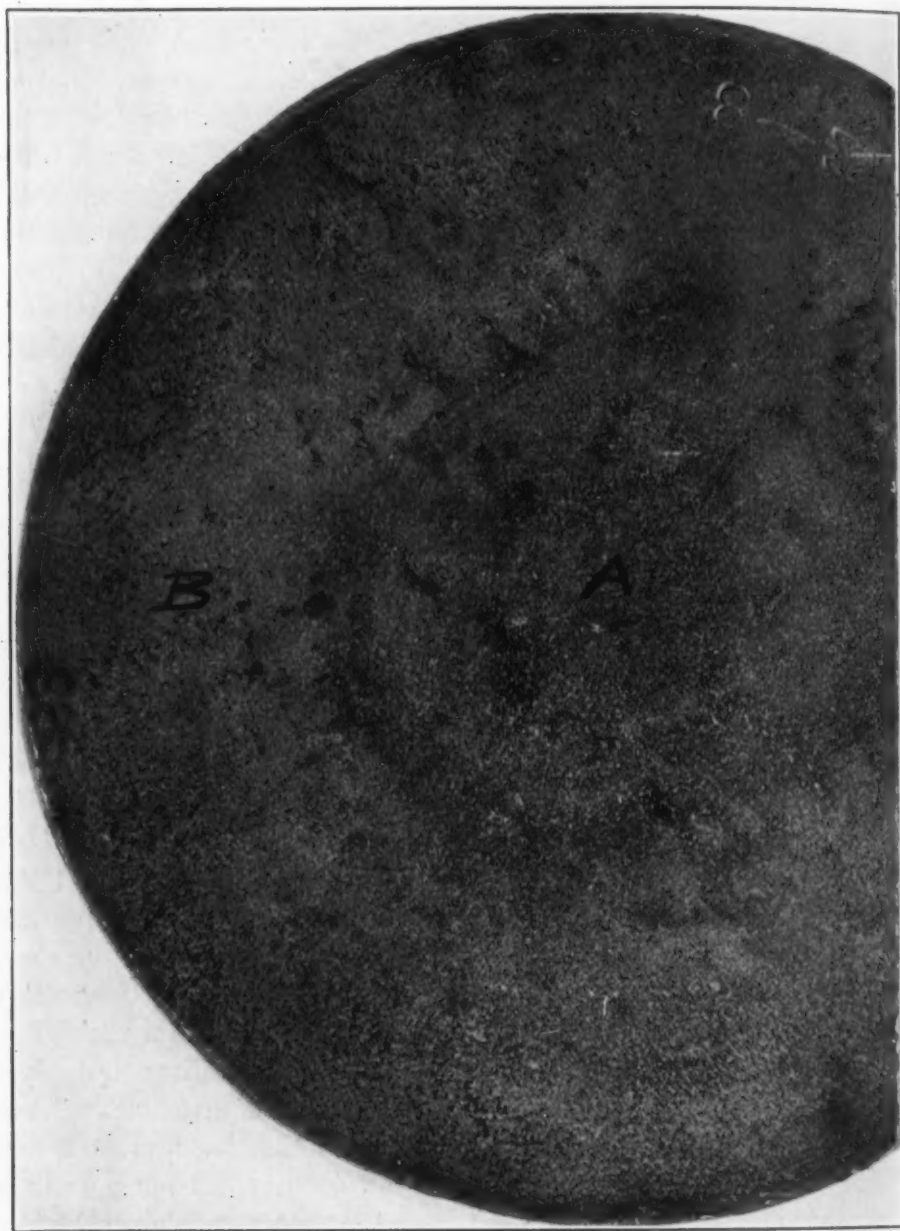Fig. 15—Segregated Open-hearth Steel Showing Absence of Dendritic Structure. (A)
—Carbon, 0.164; Manganese, 0.82; Phosphorus, 0.124; and Sulphur, 0.164. (B)—Carbon,
0.129; Manganese, 0.83; Phosphorus, 0.103; and Sulphur, 0.118.

In the preceding paragraphs we attempted to disprove the alleged
relationship between the impurities and the dendritic formation and
tried to explain the genesis of dendrites by the difference in the orien-

tations of various dendritic grains. Now in support of this view let us consider additional evidences.

The claim that it is the phosphorus that causes dendritic structure has always been attributed to Stead; but according to Stead's own statement he has never made such remarks.[13] Furthermore, he he does not consider chemical variation as being necessary for dendritic formation. Another authority, N. T. Belaiew,[4] found dendritic structure in steels free from phosphorus.

Oxygen and sulphur have also been considered as the cause of dendritic structure but according to Belaiew again, his experiments showed that steels free from these elements all developed dendritic structure. He further reasons that if the oxygen were the cause of dendrites we should get less dendritic structure in steels made by the crucible process than by other methods. Belaiew's conclusion is borne out by the writer's own observation, who has inspected many tons of steel made by the crucible method and has not noted the dendritic structure in them any less pronounced than in steels made by other methods where the melting conditions are more favorable to the absorption of oxygen. Likewise the writer has etched high sulphur and phosphorus-bearing steel which showed very little dendritic structure and in some instances, none at all. In Fig. 15 is shown an etched section of such a bar of steel which, although high in both sulphur and phosphorus and considerably segregated, is entirely free from dendrites. In Fig. 16 is also shown the structure of Armco iron which although showing considerable impurities is free from dendritic structure. The writer has also observed numerous cases of high and low carbon and alloy steels which, although badly contaminated with impurities, were free from dendrites. Therefore, we must conclude that the formation of dendrites is independent of these impurities. The carbon on the other hand seems to play a very important part in the formation of dendrites. In Fig. 16 is shown the structure of a disk of Armco iron which shows considerable segregation of impurities and is low in carbon, but free from dendrites. On the other hand Fig. 17 is the structure of a steel containing 0.36 per cent carbon made from electrolytic iron, has well developed dendrites.[14] It is evident therefore, that the presence of carbon has an important effect on the

[13]J. E. Stead, Written Discussion of J. V. W. Humfrey's paper on "Macroetching and Macroprinting," *Journal*, Iron and Steel Institute, 1919, Vol. 1, p. 292.

[14]A. Sauveur and V. N. Krivobok, "Use of Sodium Picrate in Revealing Dendritic Structure in Iron Alloys," *Transactions*, American Institute of Mining and Metallurgical Engineers, Vol. LXX, 1924, p. 239-253.
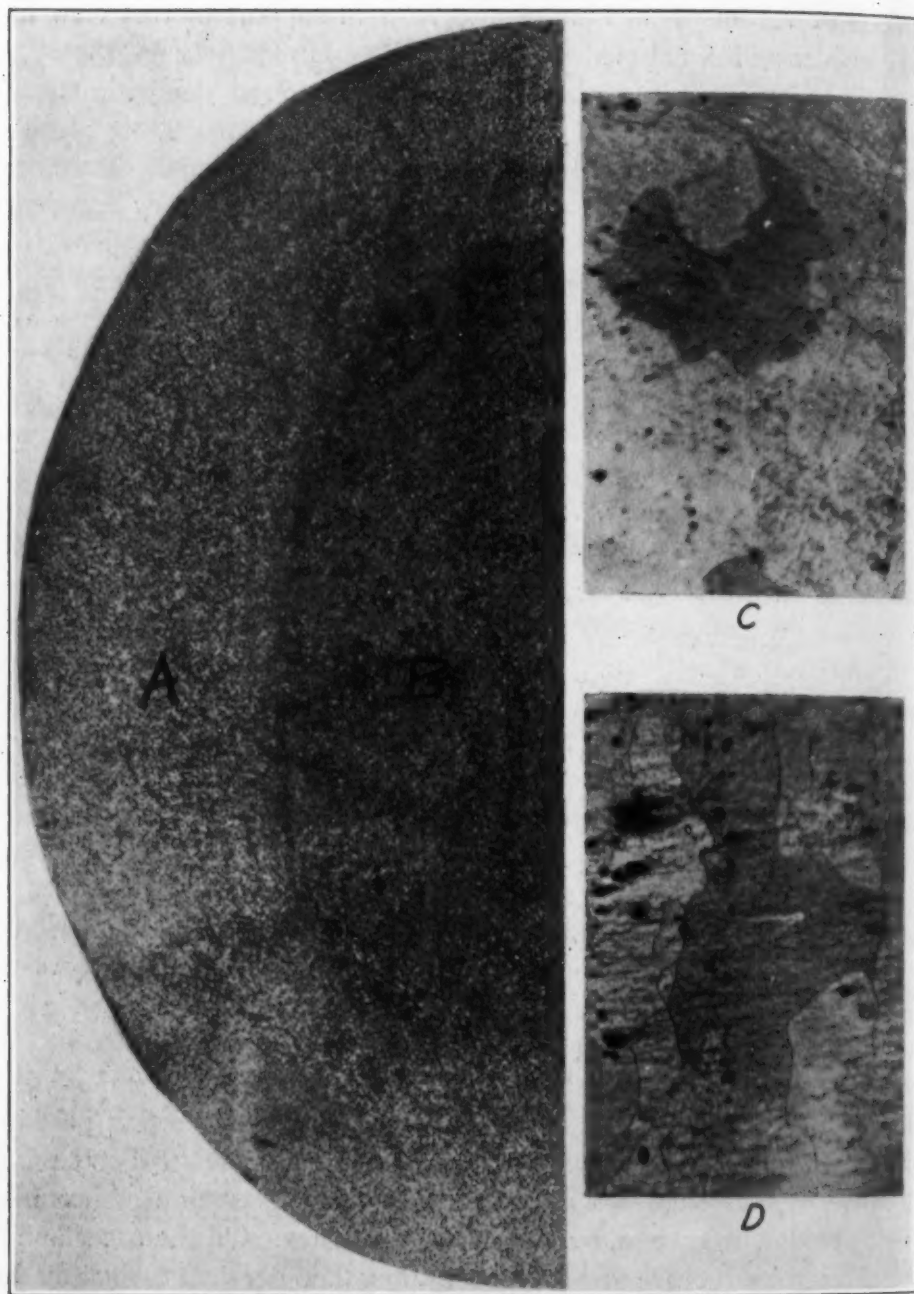
Fig. 16—Armco Iron Showing the Presence of Segregation Without Dendritic Structure. Etched in 7 Parts Hydrochloric Acid, 1 Part Sulphuric Acid, 2½ Parts Water. Actual Size. (C)—Microstructure of (A). × 100. (D)—Microstructure of (B). × 100. (A)—Carbon, 0.035; Manganese, 0.07; Silicon, 0.01; Phosphorus, 0.010 and Sulphur, 0.020. (B)—Carbon, 0.039; Manganese, 0.07; Silicon, 0.01; Phosphorus, 0.010 and Sulphur, 0.044.

formation of dendrites; besides it is a well known fact that as the carbon content increases the degree of dendritic structure in plain carbon

steels increases. The mechanism of its direct effect seems to lower the viscosity or to increase the fluidity of the molten metal and thus facilitates the formation of dendrites. In that case it becomes an intensifier rather than the generic cause of dendrites.

SEGREGATION OF IMPURITIES IN DENDRITIC STEEL

While there may not exist any causative relation between the impurities such as sulphur, phosphorus, oxygen, etc., and the den-
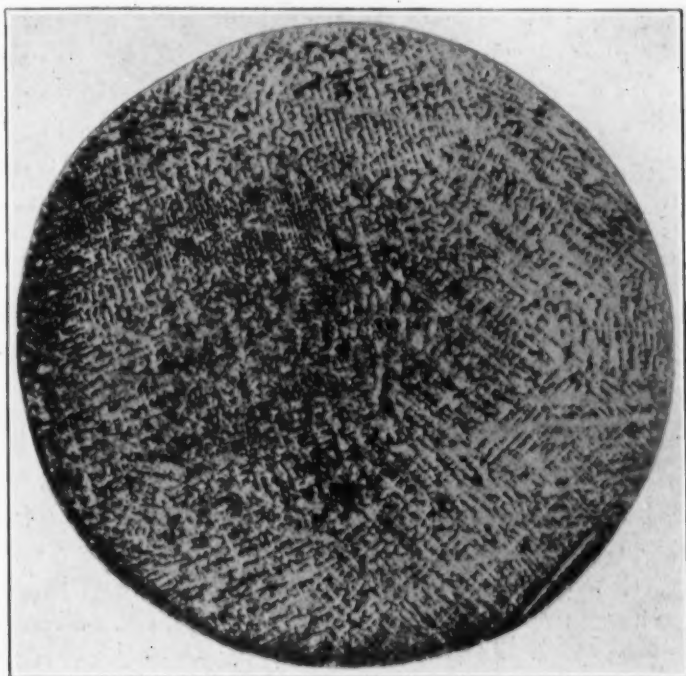
Fig. 17—Dendritic Structure in Cast Steel Containing 0.36 per cent Carbon. Made by Melting Electrolyic Iron with Sugar Charcoal in Vacua. (Sauveur and Krivobok).

drites, yet the segregation of these impurities in a dendritic steel has often been a matter of controversy. There seems to be a general tendency to believe that the segregation takes place more seriously in dendritic than nondendritic steel, which appears to be based on the fact that the fundamental factor that contributes to the formation of dendrites is slow cooling which promotes segregation. This, however, is not always true, because there are other factors beside slow cooling which promote segregation, such as, the kind and the relative amount of segregative impurities in the steel, the methods of melting and deoxidizing the molten metal, etc. In Fig. 18 are shown two steels each
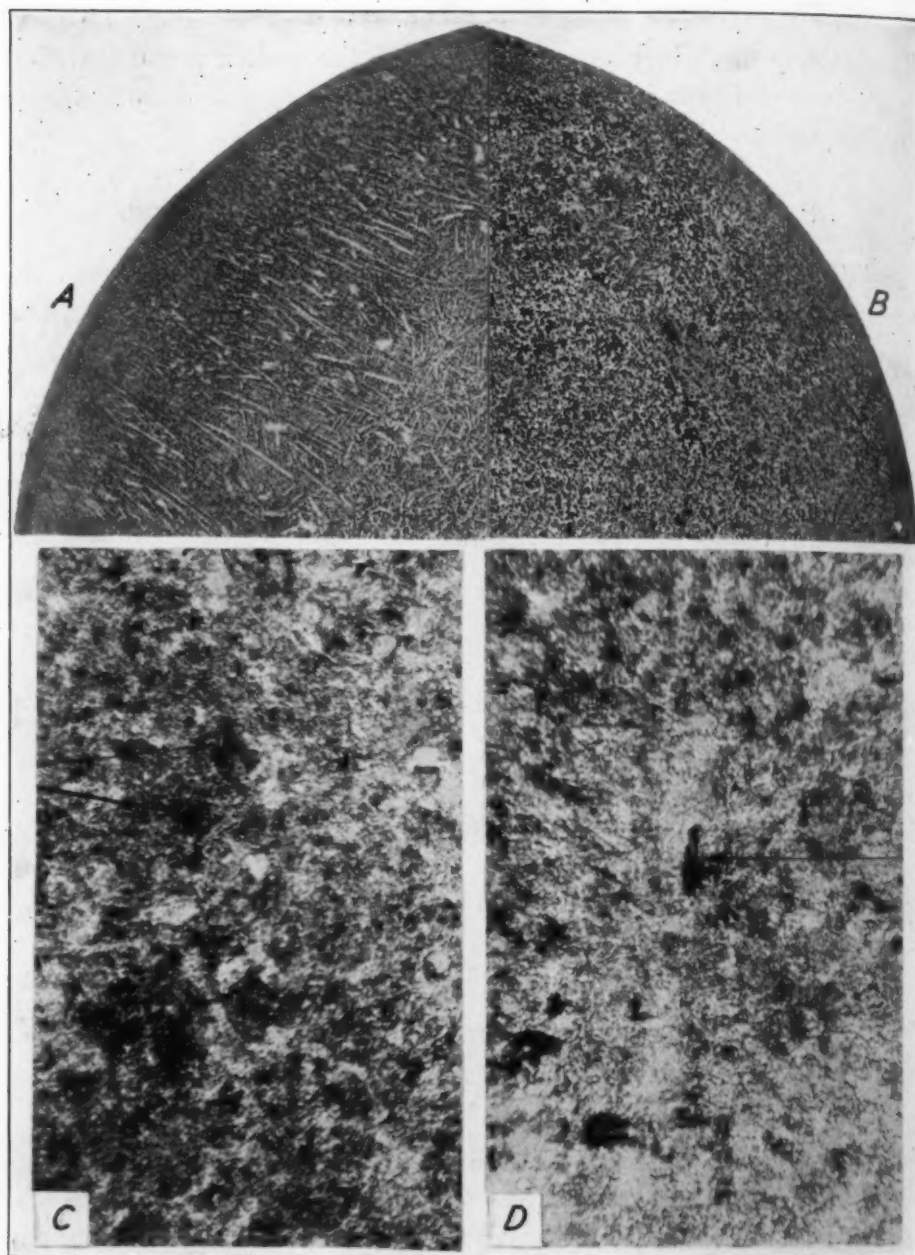
Fig. 18—(A) Macrostructure of Dendritic 1.00 per cent Carbon Tool Steel "As Received." From 5-Inch Diameter Bar Stock. Actual Size. (B) Macrostructure of Nondendritic 1.00 per cent Carbon Tool Steel "As Received." From 4½-Inch Diameter Bar Stock. Actual Size. (C) Microstructure of (A). × 100. (D) Microstructure of (B). × 100. Arrows Indicate Slag Inclusions.

containing about 1.00 per cent carbon but made in different steel mills. One is highly dendritic, the other is not, yet the impurities in the dendritic steel are not more than in nondendritic steel.
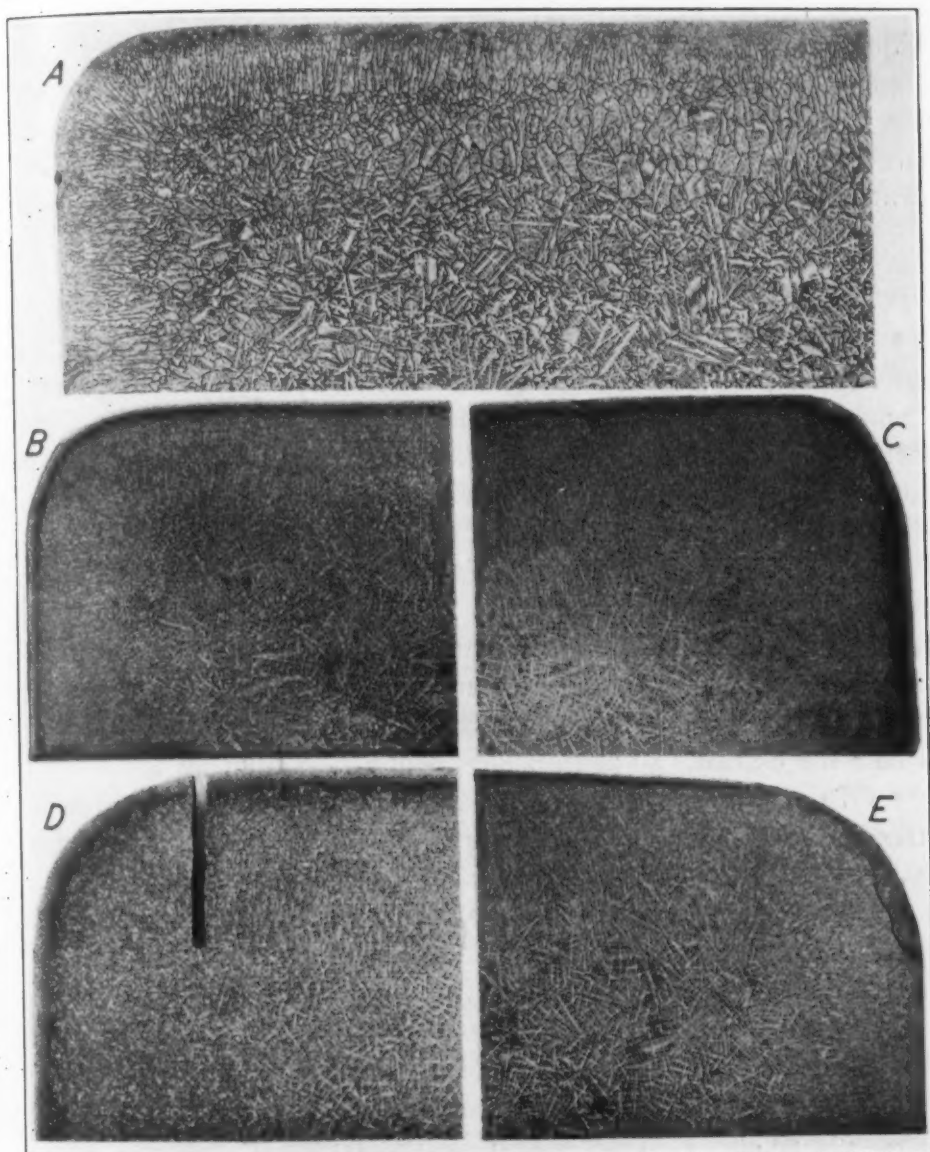
Fig. 19—Showing the Effect of Heat Treatment on Dendritic Structure. (A) Dendritic Structure of 1.00 per cent Carbon Steel Ingot Shown in Fig. 7. (B), (C), (D) and (E) Heated to 2000, 2100, 2200, and 2300 Degrees Fahr. Respectively and Held at Heat for One Hour and Slowly Cooled. All Etched in the Etching Solution Previously Mentioned. All Photographs Actual Size Except (A) which is Slightly Reduced.

Furthermore, slow cooling which promotes dendritic structure, allows more time to the impurities to agglomerate and rise to the surface or the discard end of the ingot. To a certain extent, dendritic branches mechanically entangle some of the impurities which tends to more uniformly distribute them.

In the case of less dendritic or nondendritic steel a reverse condition prevails, namely, a rapid cooling causes the impurities to be trapped in situs and prevents their agglomeration whereby they could rise to the surface. For these reasons, excepting, perhaps, the very large ingots, one would be inclined to prefer a dendritic steel to a nondendritic one if the question of impurities were only to be considered.

MECHANICAL PROPERTIES OF DENDRITIC STEEL

It can be said that there are two basic factors which control the mechanical properties of a metal, namely composition and structure. Composition indicates the kind and the relative amount of the elements which build up the metal. Structure refers to the manner in which the elements arrange themselves. In the case of steel, for instance, not only the mere presence of elements such as carbon, manganese, sulphur, phosphorus, etc., but also the way or the condition in which they are present determine the mechanical properties of the steel. The thermal and mechanical treatments which are often considered as some of the factors controlling the characteristics of the steel, are in reality nothing but artificial means to rearrange or distribute the elements so as to bring about a particular structure.

Of the two factors mentioned above we are here interested in the structure only, which can be divided into the two well known types, that is, the structures which are visible to the eye and those that are not. In other words, macroscopic and microscopic structures. A dendritic steel has both of these types of structure; but a nondendritic steel, broadly speaking, has only the microscopic structure. These are shown in Figs. 20 to 27.

In a dendritic steel the visible macrostructure consists of dendrites and the microscopic structure consists of the fine grains of the metal and usually of a network of excess ferrite or cementite as the case may be. Thus we have two structural systems in a dendritic and one structural system in nondendritic steel. Now, in order to be able to determine the effect of the dendritic structure on the mechanical characteristics of steel we must know the individual part played by each system of structures in the development of these properties, but before we can do that we must briefly bring out the characteristics of each system and their relation to each other.

In a dendritic steel there is no intimate relation between the visible dendrites and the microscopic structure or network. In other

words, in their occurrence and distribution, these two systems are independent of each other. This is a fact that has been proven by both Giolitti and Belaiew and can be shown by first photographing the macro and microstructures of a dendritic steel separately and then by making a print from the carefully superimposed negatives; or by first etching the specimen for dendritic structure, then by slightly polishing the specimen and etching it in the usual way for microstructure which will disclose the dual system of structure on the same specimen.

In addition to the independence of existence between the two systems there is also another and perhaps a most important difference between the two. This lies in their degree of responsiveness to thermal and mechanical treatments. The visible dendrites cannot be obliterated or changed by ordinary thermal and mechanical treatments but the microstructure can be. These are shown in Figs. 19 to 27.

Again, in studying the relation and characteristics between the microstructure of nondendritic and the microstructure of dendritic steel of the same analysis we find that the size of the grain and network of nondendritic steel is rather smaller than those of dendritic steel; but both are similarly responsive to thermal and mechanical heat treatment. This is shown in Fig. 25 where the network structure of dendritic and nondendritic steel have totally altered after the heat treatment. After noting the characteristics of these structural systems and the relation existing among them, let us now attempt to determine the influence which each system exerts on the mechanical properties of the steel.

There seems to be two ways of attacking the problem. (1)—By taking a dendritic steel and eliminating the dendritic structure by a series of drastic heat treatments as the one used by Giolitti or by a combination of heat treatment and hot work and then conducting mechanical tests on specimens which have received various degrees of these treatments and then comparing the test results. A test of this nature would no doubt show an increase in mechanical properties from the original cast dendritic steel to the stage where the dendrites have been eliminated or nearly so; but it will not show the part played by each individual system of structure; because these increased properties may result from the refining effect of the treatments on the microstructure as well as from the elimination of dendrites. Hence, it cannot be told with sufficient clarity how much of the improved characteristics is due to the elimination of dendrites and how much to the improved state of microstructure.

(2)—A second method which the author has used consisted of taking a dendritic and a nondendritic steel of the same analysis and carrying out various tests after identical thermal and mechanical treatments. For this purpose the dendritic and nondendritic hypereutectoid carbon steel ingots shown in Figs. 7 and 8 the dendritic and non-dendritic hypoeutectoid chromium-nickel steel ingots shown in Figs.

Table I

Physical Properties of Dendritic and Nondendritic Hypoeutectoid Chromium-Nickel Steel Ingot S.A.E. 3312 Annealed at 1350 Degrees Fahr. Fig. 20.

Mark	Tensile Strength Lbs. Per Sq. Inch	Elastic Limit Lbs. Per Sq. Inch	Elongation In 2 In. Per Cent	Reduction of Area Per Cent	Brinell Hardness Number
Dendritic					
G-88	54,100	.....	...	.. ⎫	
G-89	82,100	.....	...	.. ⎬	228
G-90	76,900	.....	...	.. ⎭	
Average	71,000				
Nondendritic					
H-84	63,700	.....	...	.. ⎫	
H-85	51,000	.....	...	.. ⎬	228
Average	57,300				

9 and 10 were used. Specimens were tested, (a) after annealing below the critical, (b) after normalizing and annealing, (c) after forging to the same reduction of area, then normalizing and annealing.

Tables I, II and III show the relative mechanical properties of dendritic and nondendritic low carbon, chromium-nickel, steel S.A.E. 3312, after different thermal and mechanical treatments. Tables IV, V and VI show the relative mechanical properties of dendritic and nondendritic 1.00 per cent carbon steel after different thermal and mechanical treatments. Table VII shows the relative physical properties of dendritic and nondendritic low carbon-nickel steel S.A.E. 2320 in as received condition. Table VIII shows the relative directional properties of dendritic and nondendritic S.A.E. 3312 steel given above after 5 to 1 reduction of area and heat treatment. Table IX shows the relative fatigue resisting properties of 1.00 per cent carbon dendritic and nondendritic steel after forging to 5 to 1 reduction of area and normalizing and annealing treatment. Table X shows the relative wearing properties of dendritic and nondendritic carbon tool steels tested for drawing brass and copper tubing.

From Table I it is to be noted that both specimens of this composition are quite brittle, but the dendritic steel has greater tensile strength.

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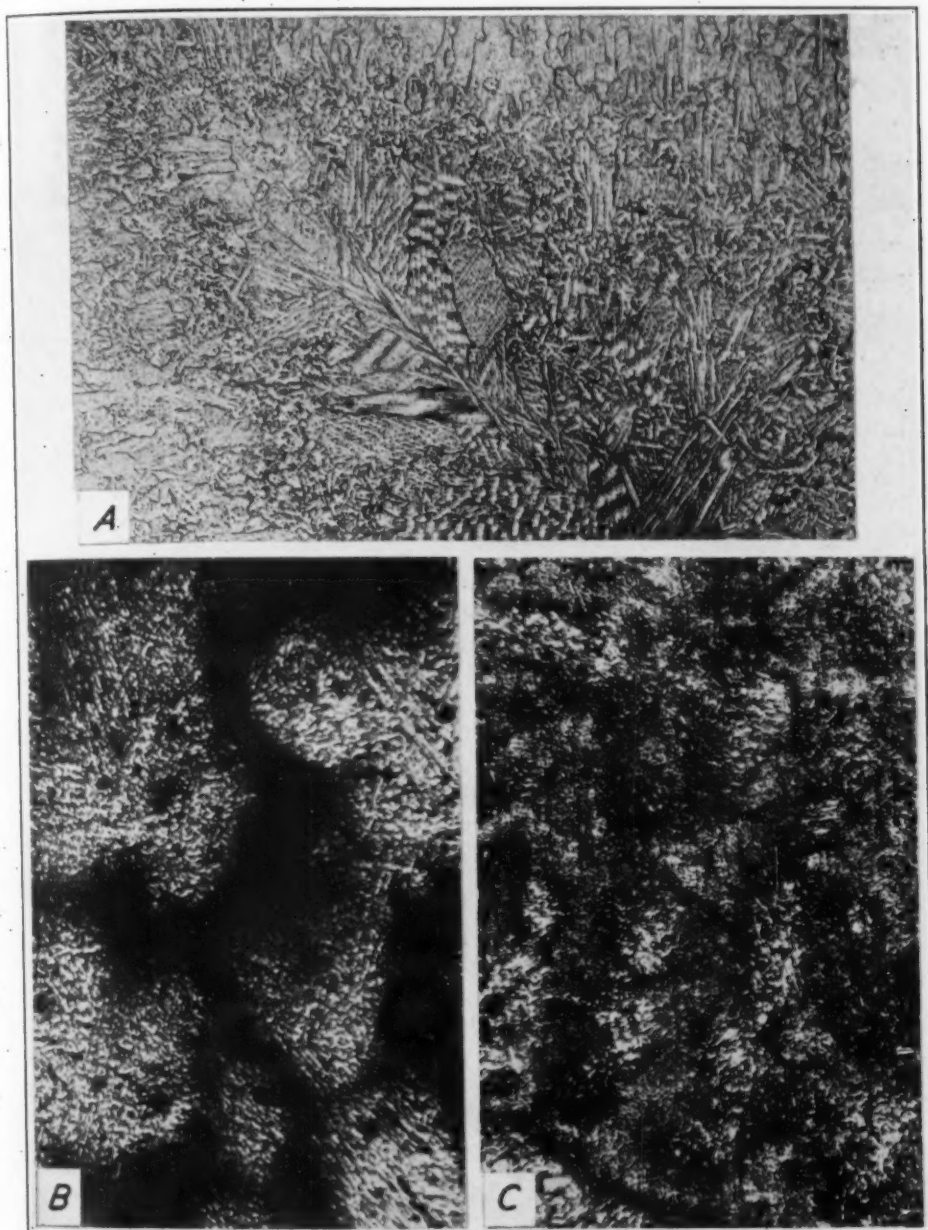
Fig. 20—(A) Macrostructure of 6 by 6-Inch Dendritic Chromium-Nickel Steel Ingot Shown in Fig. 9, S.A.E. 3312, As Cast. Actual Size. (B) Microstructure of Same. × 100. (C) Microstructure of Steel (A) After Oil Quenching from 1650 Degrees Fahr. and Tempering at 1300 Degrees Fahr. × 100.

From Table II it will be noted that both specimens show improved properties after normalizing heat treatment, but dendritic specimens still show better physical properties. This perhaps is due
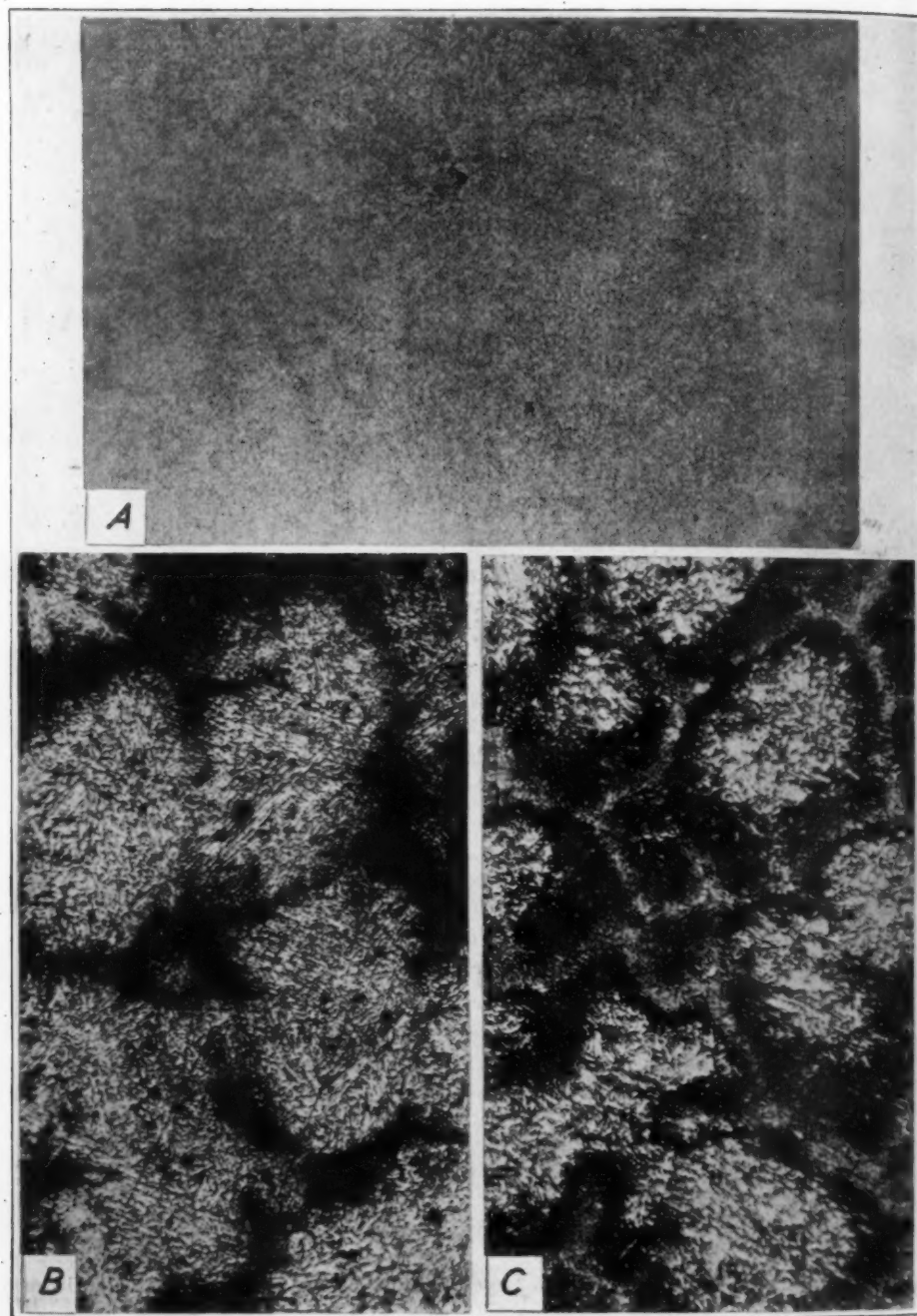
Fig. 21—(A) Macrostructure of Nondendritic Chromium-Nickel Steel, S.A.E. 3312, Shown in Fig. 10. "As Cast" and Actual Size. (B) Microstructure of Same. × 125. (C) Steel (A) Oil-Quenched from 1650 Degrees Fahr. and Tempered at 1300 Degrees Fahr. × 100.

Fig. 22—(A) Macrostructure of 6 by 6-Inch Dendritic Chromium-Nickel Steel, S.A.E. 3312; Shown in Fig. 9 after Forging to 5 to 1 Area Reduction and Oil Quenching from 1650 Degrees Fahr. and Tempering at 1300 Degrees Fahr. Actual Size. (B) Microstructure of Same. × 100. (C) Macrostructure of 4 by 4-Inch Nondendritic Chromium-nickel Steel, S.A.E. 3312, after Forging to 5 to 1 Area Reduction and Oil Quenching from 1650 Degrees Fahr. and Tempering at 1300 Degrees Fahr. Actual Size. (D) Microstructure of Same. × 100.

Table II

Physical Properties of Dendritic and Nondendritic Hypoeutectoid Chromium-Nickel Steel S.A.E. 3312, Oil-quenched from 1650 Degrees Fahr. and Tempered at 1300 Degrees Fahr. Fig. 21.

Mark	Tensile Strength Lbs. Per Sq. Inch	Yield Point Lbs. Per Sq. Inch	Elongation In 2 In. Per Cent	Reduction of Area Per Cent	Brinell Hardness Number
Dendritic					
I-96	125,000	88,500	4.00	2.7	
I-97	133,000	37,300	4.00	2.7	269
I-98	137,000	89,000	6.00	8.1	
Aver.	131,600	88,200	4.60	4.5	
Nondendritic					
J-92	89,400	77,400	2.50	1.00	
J-93	114,000	79,000	2.50	2.70	269
Aver.	101,700	78,200	2.50	1.85	

Table III

Physical Properties of Dendritic and Nondendritic Hypoeutectoid, Chromium-Nickel Steel S.A.E. 3312, after Forging to 5 to 1 Reduction of Area and Oil Quenching from 1650 Degrees Fahr. and Tempering at 1300 Degrees Fahr. Fig. 22.

Mark	Tensile Strength Lbs. Per Sq. Inch	Yield Point Lbs. Per Sq. Inch	Elongation In 2 In. Per Cent	Reduction of Area Per Cent	Brinell Hardness Number
Dendritic					
K-48	144,200	96,000	10.00	26.0	
K-49	145,600	97,200	11.00	26.0	
K-50	145,600	94,000	11.00	26.1	321
K-51	144,500	97,200	10.00	26.1	
Aver.	145,000	96,100	10.50	26.05	
Nondendritic					
L-56	143,600	79,400	10	28.0	
L-57	139,200	99,600	12	32.0	321
L-59	143,000	130,000	10	20.4	
Aver.	141,900	102,500	10.6	26.8	

Table IV

Relative Physical Properties of Dendritic and Nondendritic One Per Cent Carbon Steel Ingots Annealed at 1350 Degrees Fahr. Fig. 23.

Mark	Tensile Strength Lbs. Per Sq. Inch	Elastic Limit Lbs. Per Sq. Inch	Elongation In 2 In. Per Cent	Reduction of Area Per Cent	Brinell Hardness Number
Dendritic					
A-64	62,000	......	1.00	4.7	
A-65	55,000	......	....	...	269
A-66	58,000	......	1.00	1.4	
Aver.	58,300	......	0.66	2.0	
Nondendritic					
B-60	113,000	71,000	0.50	1.4	
B-61	82,000	81,000	0.50	4.0	286
B-62	87,000	63,000	0.66	1.4	
Aver.	94,000	71,600	0.55	2.2	

to the smaller size of ferrite grains in dendritic sample as shown in Figs. 20C and 21C.

From Table III it will be noted that both dendritic and nonden-
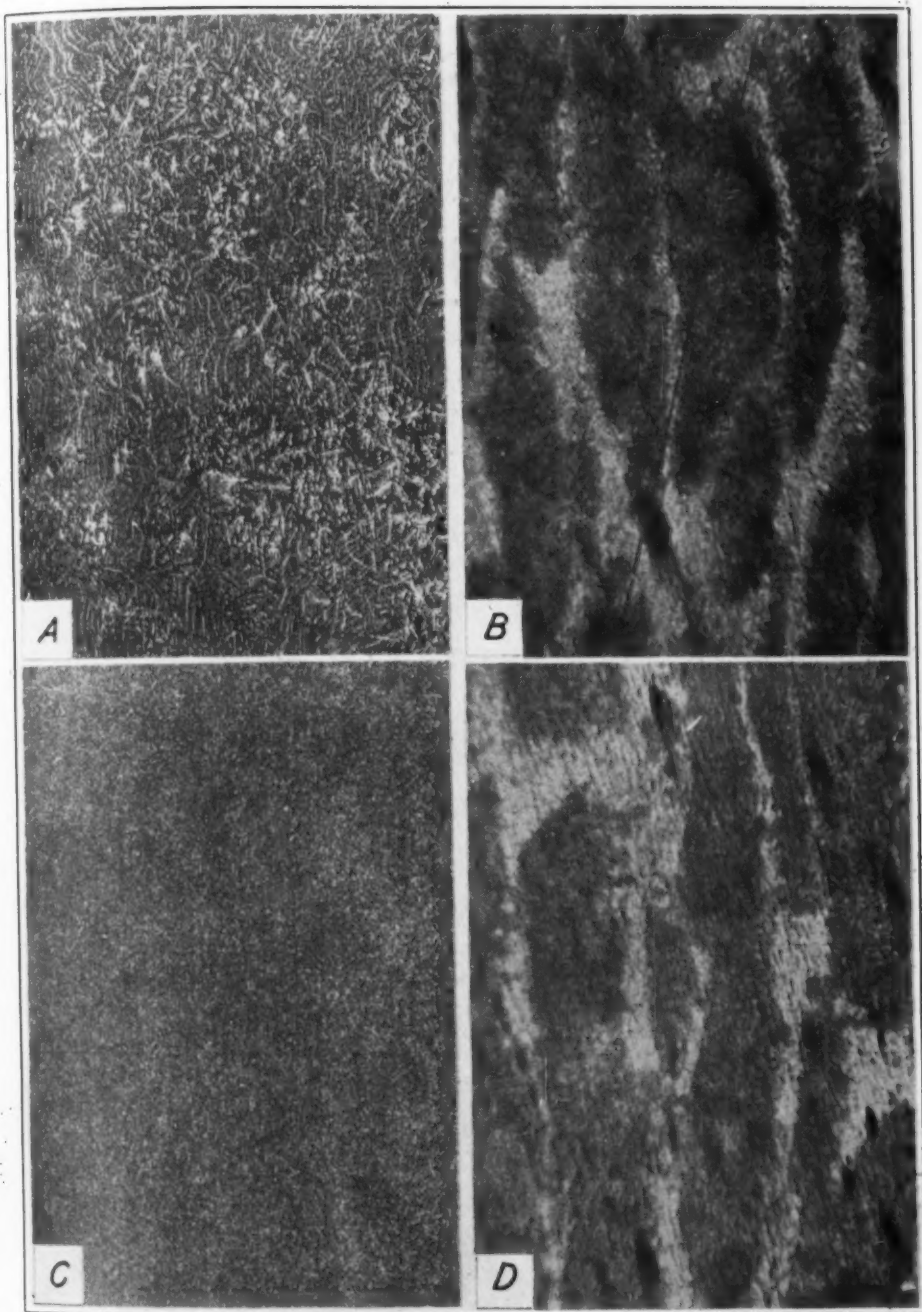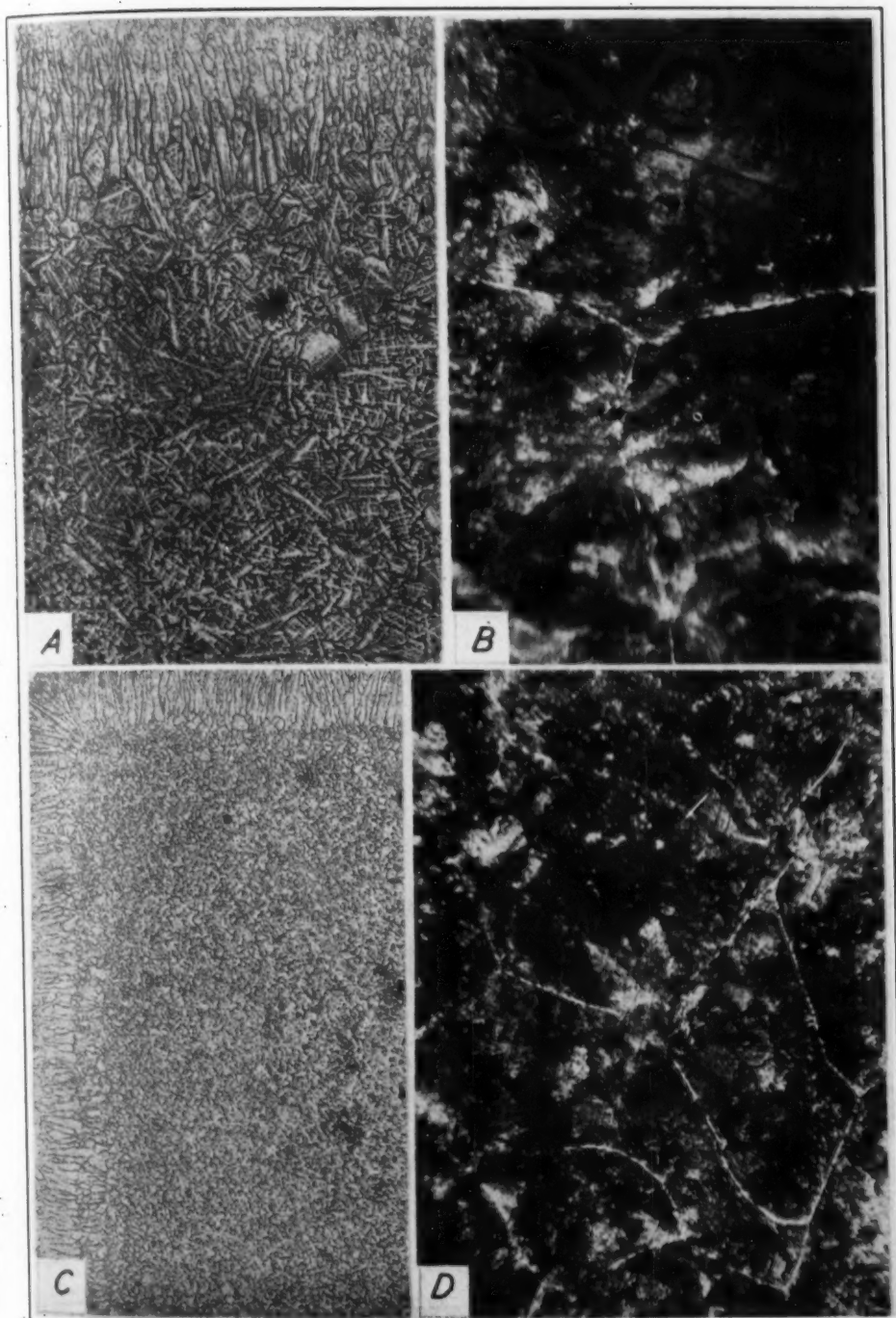
Fig. 23—(A) Macrostructure of 6 by 6-Inch Dendritic Steel Ingot, 1.00-per cent Carbon. Actual Size. (B) Microstructure of Steel (A). × 100. Note Large Network of Excess Cementite. (C) Macrostructure of 4 by 4-Inch Nondendritic 1.00-per cent Carbon Steel Ingot. Actual Size. (D) Microstructure of Steel (C). × 100. Note the Relatively Small Size of Excess Cementite.
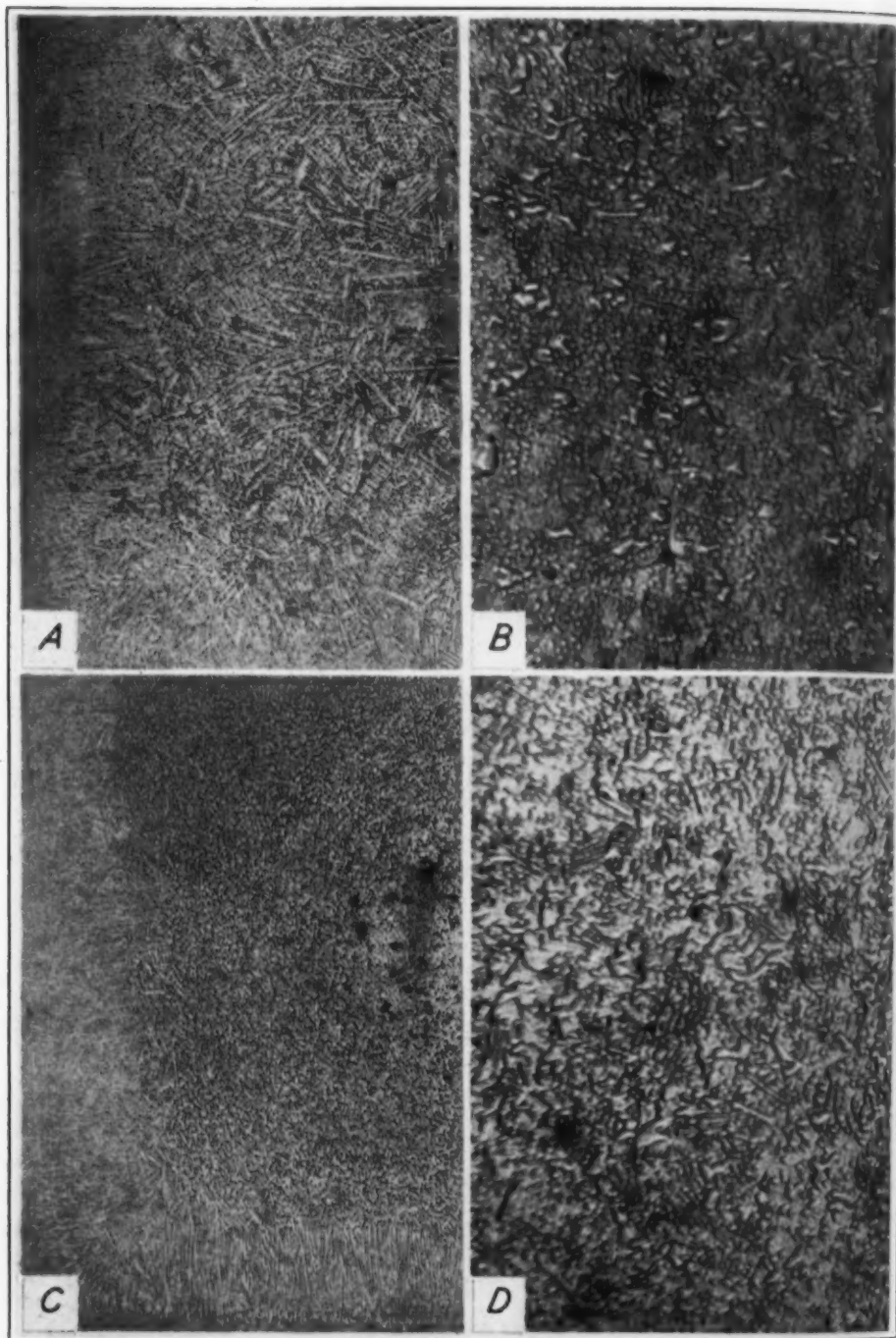
Fig. 24—(A) Macrostructure of 6 by 6-Inch 1.00-per cent Carbon Steel Ingot after Normalizing at 1650 Degrees Fahr. and Annealing at 1425 Degrees Fahr. Actual Size. (B) Microstructure of (A). × 400. (C) Macrostructure of 4 by 4-Inch 1.00-per cent Carbon Nondendritic Steel Ingot after Normalizing at 1650 Degrees Fahr. and Annealing at 1425 Degrees Fahr. Actual Size. (D) Microstructure of (C). × 400.
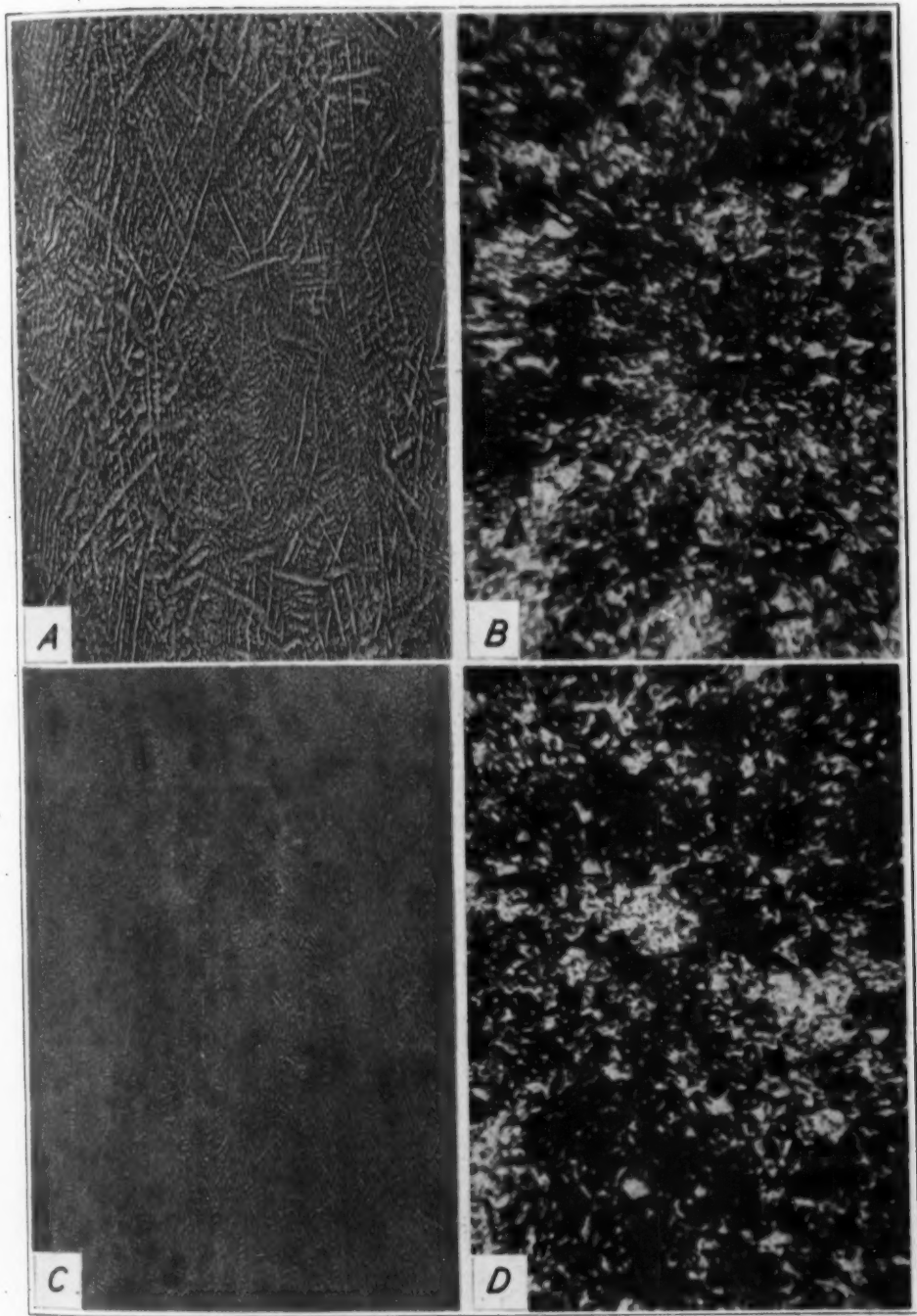
Fig. 25—(A) Macrostructure of 6 by 6-Inch Dendritic 1.00-per cent Carbon Steel Ingot after Forging to 4 to 1 Area Reduction and Normalizing at 1650 Degrees Fahr. and Annealing at 1425 Degrees Fahr. Actual Size. (B) Microstructure of Same. × 400. (C) Macrostructure of 4 by 4-Inch Nondendritic 1.00-per cent Carbon Steel Ingot After Forging to 4 to 1 Area Reduction and Normalizing at 1650 Degrees Fahr. and Annealing at 1425 Degrees Fahr. Actual Size. (D) Microstructure of Same. × 400.

Table V

Relative Physical Properties of Dendritic and Nondendritic 1.00 Per Cent Carbon Steel Ingots after Normalizing at 1650 Degrees Fahr. and Annealing at 1450 Degrees Fahr. Fig. 24.

| Mark	Tensile Strength Lbs. Per Sq. Inch	Elastic Limit Lbs. Per Sq. Inch	Elongation In 2 In. Per Cent	Reduction of Area Per Cent	Brinell Hardness Number
Dendritic					
C-72	76,200	47,000	5:00	3.6 ⎫	
C-73	74,300	46,000	7.50	6.3 ⎬ 187	
C-74	79,700	47,000	6.00	6.3 ⎭	
Aver.	76,700	46,600	6.16	5.4	
Nondendritic					
D-68	86,200	47,200	19.00	23.60 ⎫	
D-69	86,300	48,700	17.50	25.40 ⎬ 187	
D-70	87,000	49,500	19.00	24.50 ⎭	
Aver.	86,500	48,400	18.50	24.50	

dritic steels have practically the same physical properties and from Fig. 22 it is to be particularly noted that both steels also have the same microstructure, although the dendritic structure still continues to exist.

In these three tables were given the relative physical properties of hypoeutectoid dendritic and nondendritic chromium-nickel steel. In Tables IV to VIII are given the relative physical properties of hypereutectoid dendritic and nondendritic carbon tool steel.

From Table IV it will be seen that nondendritic steel has greater tensile strength than the dendritic steel.

From Table V it will be seen that both dendritic and nondendritic specimens after normalizing and annealing have better physical properties than the specimens in the previous test. It is to be noted, however, that the nondendritic specimens are still in the lead. These in-

Table VI

Relative Physical Properties of Dendritic and Nondendritic Carbon Steel Ingots after Forging to 4 to 1 Area Reduction and then Normalizing at 1650 Degrees Fahr. and Annealing at 1450 Degrees Fahr. Fig. 25.

Mark	Tensile Strength Lbs. Per Sq. Inch	Elastic Limit Lbs. Per Sq. Inch	Elongation In 2 In. Per Cent	Reduction of Area Per Cent	Brinell Hardness Number
Dendritic					
E-80	85,000	46,700	23.50	50.5 ⎫	
E-81	84,000	48,400	20.5	50.5 ⎬ 187	
E-82	85,000	48,400	26.5	52.50 ⎭	
Aver.	84,600	47,800	23.5	51.1	
Nondendritic					
F-76	85,000	48,000	22.5	46.3 ⎫	
F-77	84,700	46,700	21.5	43.5 ⎬ 187	
F-78	84,800	47,700	23.0	46.3 ⎭	
Aver.	84,800	47,200	22.3	45.3	

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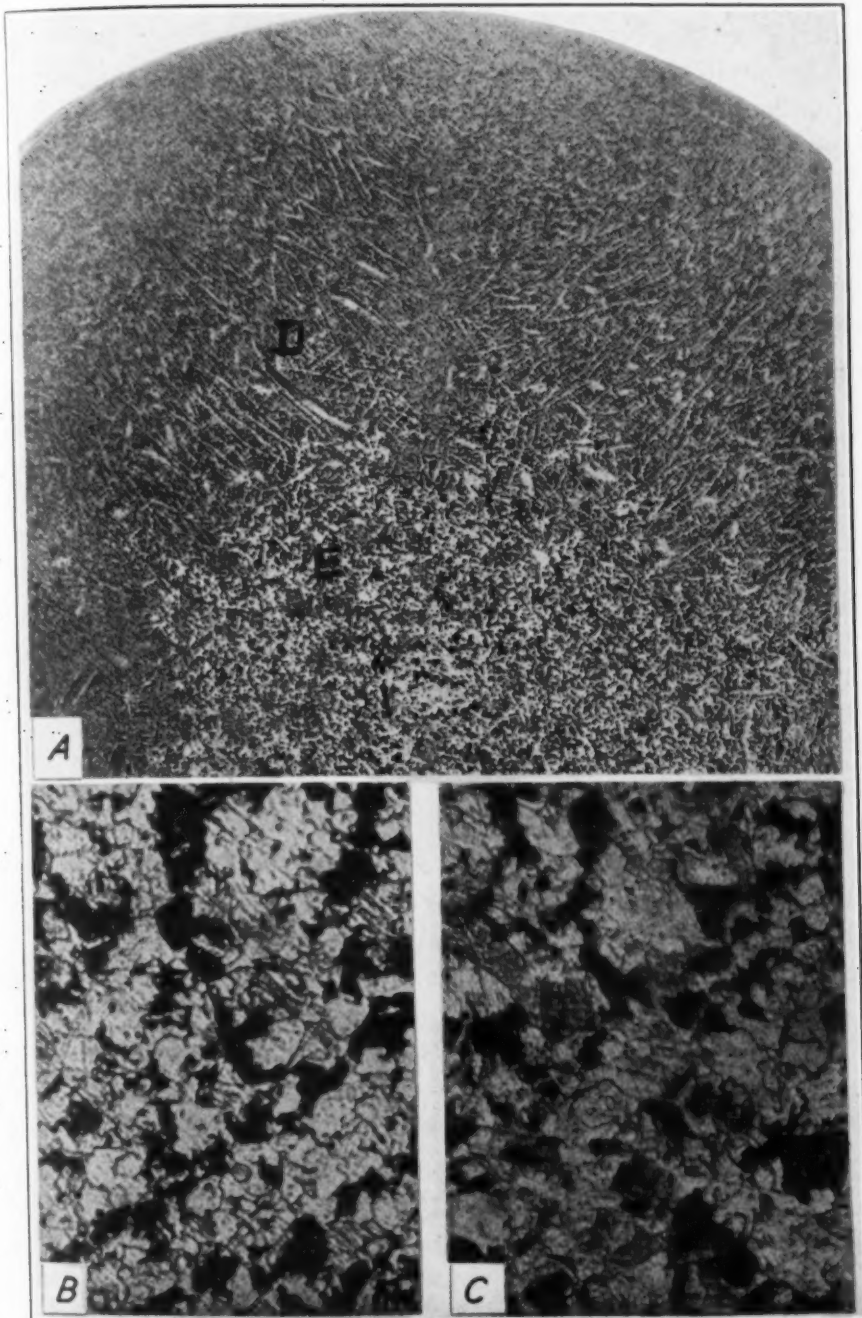Fig. 26—(A) Macrostructure of 4½-Inch Diameter S.A.E. 2320 Nickel Steel "As Received" Showing a Dendritic Outside and a Nondendritic Inside. (B) Microstructure of Outside. (C) Microstructure of Inside. Both × 300. (D) Carbon, 0.177; Manganese, 0.69; Phosphorus, 0.023; Sulphur, 0.012; Chromium, 0.32 and Nickel, 3.39. (E) Carbon, 0.181; Manganese, 0.69; Phosphorus, 0.021; Sulphur, 0.019; Chromium, 0.24 and Nickel, 3.36.

creased properties clearly emphasize the importance of normalizing heat treatments of cast steel parts in practice and they particularly show the desirability of securing a nondendritic structure in such castings which when properly heat treated will give greatly improved properties.

The comparison of results in Table VI shows that both dendritic and nondendritic steels have nearly the same physical properties.

Table VII

Relative Physical Properties of Dendritic Outside and Nondendritic Center of Nickel Steel Shown in Fig. 26. Tested in "as-received" condition.

Mark	Tensile Strength Lbs. Per Sq. In.	Elastic Limit Lbs. Per Sq. In.	Elongation in 2 Ins. Per Cent	Reduction of Area Per Cent
Outside or Dendritic				
0-1	79,000	52,000	32.50	66.0
0-2	78,500	......	34.50	66.0
0-3	79,500	......	35.00	64.0
0-4	78,000	51,000	33.00	66.0
Aver.	78,700	51,500	33.25	65.50
Inside or Nondendritic				
C-1	78,500	54,000	32.00	66.0
C-2	78,000	55,500	31.50	64.0
C-3	79,000	53,000	30.50	64.0
C-4	78,000	50,500	32.50	64.0
Aver.	78,300	53,200	31.62	64.5

Furthermore, dendritic steel has somewhat higher reduction of area than the nondendritic steel.

Again from Fig. 25 it will also be seen that the dendritic structure after forging, although quite well deformed, still continues its existence, but it no longer seems to interfere with the free functioning of the microstructure of the metal and appears to drop out of the picture as affecting the physical properties of the steel, at least the properties indicated here.

Perhaps before going further, it would be well to illustrate by a concrete example one of the conclusions of practical value which can be drawn from the preceding tests, that is, in the necessity of passing judgment on the physical properties of a dendritic steel the thermal and mechanical treatments which the metal might have received in the steel mill must not be forgotten.

For instance in Fig. 26 is shown the microstructure of a bar of S.A.E. 2320 nickel steel as received from the steel mill. It shows a well pronounced dendritic structure on the outside and a nondendritic structure in the center, which would lead one to the thought that the steel has different physical properties in the two different areas; but the results of physical tests made on samples cut out from the two

areas and parallel to the axis of the bar show that this is not the case and the thermal and mechanical treatments which it had received in the steel mill have eliminated the effect of dendrites, as shown in Table VII.

DIRECTIONAL PROPERTIES OF DENDRITIC STEEL

It is often stated that a dendritic steel after thermal and mechanical heat treatment will show considerable difference in its directional properties, when tested longitudinally and transversely due to the fibrous structure resulting from the elongation of dendrites and impurities along the forging or rolling direction. To determine the extent of this difference the following test was made. For this purpose the low carbon chromium-nickel steel ingots S.A.E. 3312 steel shown in Figs. 9 and 10 were used. They were forged to 5 to 1 area reduction, then oil-quenched from 1650 degrees Fahr. tempered to 321 Brinell hardness. Both dendritic and nondendritic specimens were tested longitudinally and transversely. The structures of both steels

Table VIII

Directional Properties of Dendritic and Nondendritic Chromium-Nickel Steels shown in Fig. 22.

DENDRITIC

Mark	Tensile Strength Lbs. Per Sq. Inch	Elastic Limit Lbs. Per Sq. Inch	Elongation In 2 In. Per Cent	Reduction of Area Per Cent	Brinell Hardness Number
Tested Transversely					
K-44	146,400	86,800	9.0	18.1	321
K-45	155,000	98,200	9.0	19.0	321
K-46	150,000	96,000	10.0	22.4	321
K-47	151,200	97,200	10.0	25.3	321
Aver.	150,600	94,500	9.5	21.2	
Tested Longitudinally					
K-48	144,200	96,000	10.0	26.	321
K-49	145,600	97,200	11.0	26.	321
K-50	145,600	94,000	11.0	26.1	321
K-51	144,500	97,200	10.0	26.1	321
Aver.	145,000	96,100	10.50	26.05	

NONDENDRITIC

Mark	Tensile Strength Lbs. Per Sq. Inch	Elastic Limit Lbs. Per Sq. Inch	Elongation In 2 In. Per Cent	Reduction of Area Per Cent	Brinell Hardness Number
Tested Transversely					
K-52	143,000	94,000	11.0	26.1	
K-53	133,000	94,000	9.0	23.4	
K-54	146,800	98,600	9.0	19.0	321
K-55	159,800	97,800	12.0	30.0	
Aver.	145,600	96,100	10.2	24.6	
Tested Longitudinally					
K-56	143,600	79,400	10.0	28.0	
K-57	139,200	99,600	12.0	32.0	
K-59	148,000	130,000	10.0	24.0	321
Aver.	143,600	102,600	10.6	26.8	

after the above thermal and mechanical treatments are shown in Fig. 22 and the results of the test in Table VIII.

From the results in Table VIII it will be seen that (a) both dendritic and nondendritic chromium-nickel steels of the type under consideration when tested longitudinally show almost the same physical properties, although the dendritic structure still continues to exist; (b) that there is a slight difference in the directional properties of dendritic steel particularly in the reduction of area, but there is also similar difference between the directional properties of nondendritic steel; (c) in both steels the reduction of area is greater with the longitudinal than with the transverse test, being about 23 per cent for the dendritic and about 9 for the nondendritic steel; (d) that the transverse properties of dendritic and nondendritic specimens are about the same except the reduction of area in which nondendritic steel has about 16 per cent more than the dendritic.

RELATIVE FATIGUE RESISTING PROPERTIES OF DENDRITIC AND NONDENDRITIC STEEL

We have so far discussed the physical properties of dendritic and nondendritic steels; it is obvious that there are many other mechanical properties which are similarly important in practice; like compressive strength, resistance to wear, resistance to fatigue, cutting properties, etc., to deal with, each of which is altogether too large a subject for this paper. The author, however, has some data on resistance to fatigue and to wear which are given below.

Part of the dendritic and nondendritic 1.00 per cent carbon steel ingots were forged to 5 to 1 area reduction, normalized from 1650 degrees Fahr. and annealed and four specimens of each were tested on a Stanton repeated impact testing machine. The macro and microstructure of the two steels are shown in Fig. 27 and the results are given in Table IX.

It will be noted that the fatigue resisting properties of dendritic steel at an average, is about 16 per cent less than that of nondendritic steel. The difference between the individual specimens, however, is not uniformly so.

RELATIVE WEARING PROPERTIES OF DENDRITIC AND NONDENDRITIC STEEL

In order to compare the wearing properties of dendritic and non-
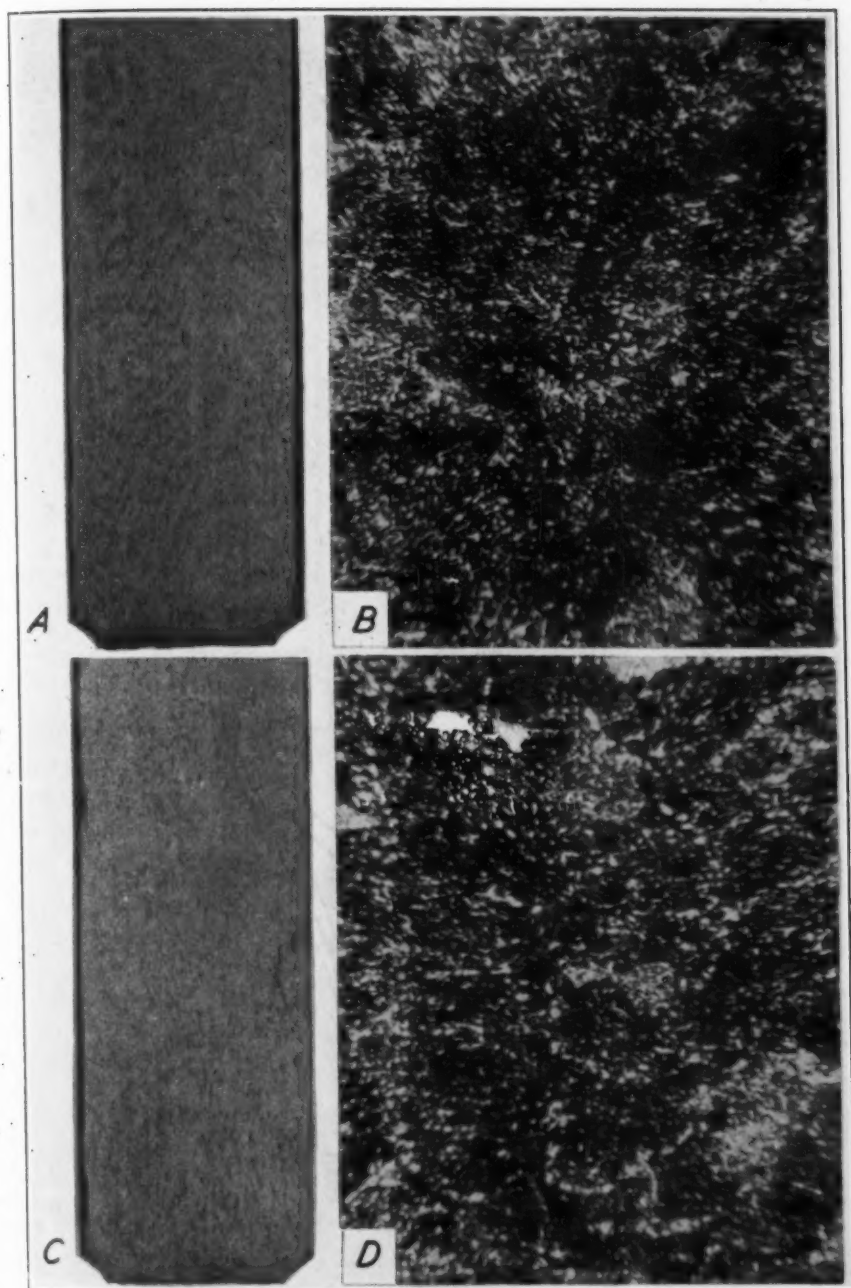
Fig. 27—(A) Macrostructure of 6 by 6-Inch Dendritic 1.00-per cent Carbon Steel Ingot after Forging to 5 to 1 Reduction and Normalizing and Annealing at 1650 and 1425 Degrees Fahr. Respectively. × 2. (B) Microstructure of Same. × 400. (C) Macrostructure of 4 by 4-Inch Nondendritic 1.00-per cent Carbon Steel Ingot after Forging to 5 to 1 Reduction and Normalizing from 1650 Degrees Fahr. and Annealing at 1425 Degrees Fahr. × 2. (D) Microstructure of Same. × 400. (A) and (C) Etched with Ammonium Persulphate.
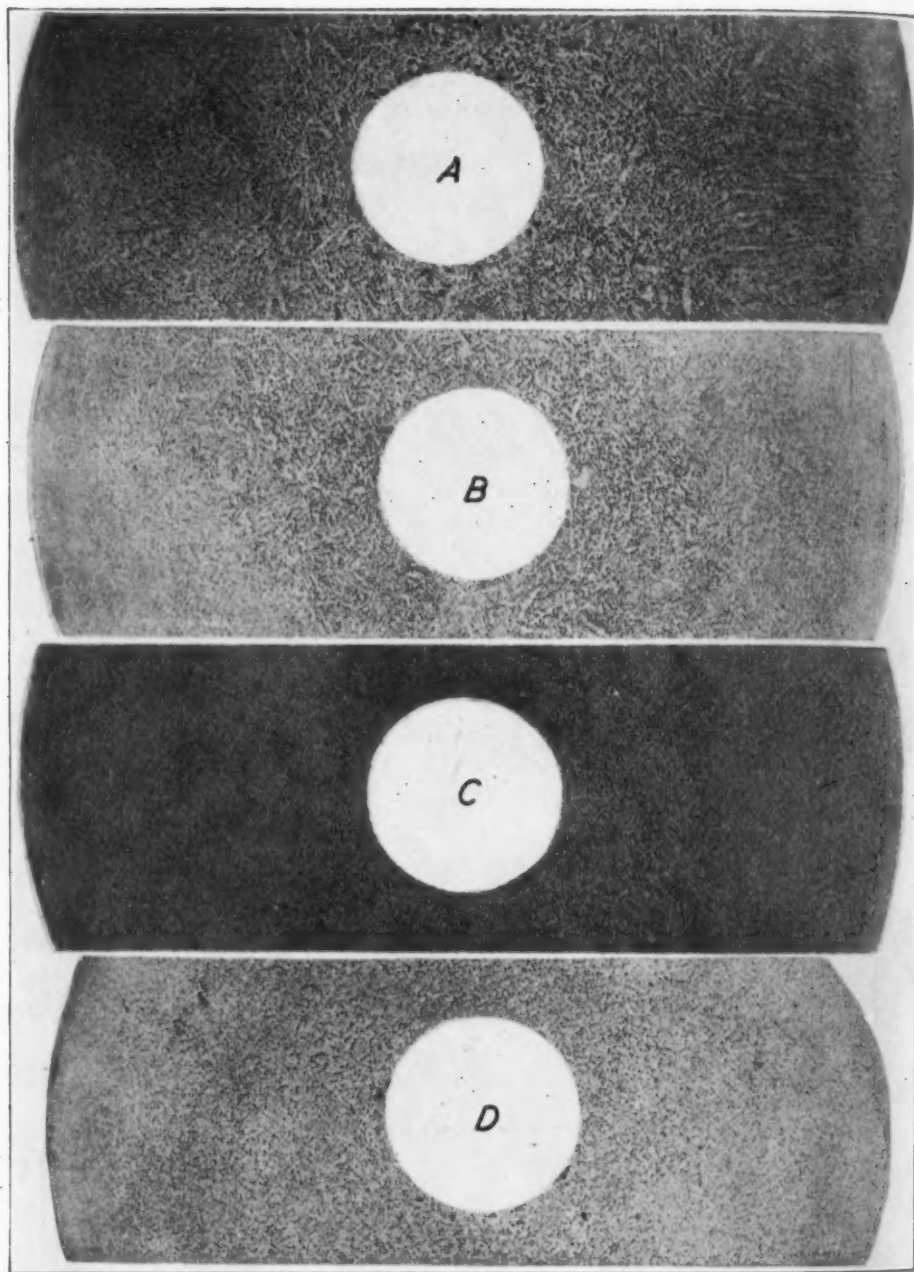
Fig. 28—Photographs of 4-Inch Diameter Dendritic and Nondendritic Carbon Tool Steel Blanks Made into Drawing Dies to Test their Relative Wearing Properties Measured in Number of Feet of Brass Condenser Tubing Drawn per 0.001 Inch Wear in the Diameter of the Die.   See Table X.

dendritic steels a number of drawing dies were made from bars of dendritic and nondendritic steels for drawing brass condenser tubing.

Table IX

Stanton Repeated Impact Tests on Dendritic and Nondendritic 1.00 Per Cent Carbon
Steel, Forged to 5 to 1, Normalized from 1650 Degrees Fahr., and Annealed
to Brinell Hardness 187.   Height of Fall of Hammer, 1.5 Inches.

Specimen	Repetition of Strokes
Dendritic	
T-1	163
T-2	137
T-3	94
T-4	115
Average	127
Nondendritic	
R-1	165
R-2	199
R-3	130
R-4	113
Average	151

Each die was rehardened and re-used three times drawing a total of
225,000 feet of tubing.

This method of testing the wearing properties of steel is not, of
course, as accurate as other wear tests of more precise nature. It is
believed, however, that as these tubes are manufactured under strict
specifications as to analysis, hardness, physical properties, grain size,
etc., the test can be considered as carried out under fairly uniform
conditions on a uniform material.

The typical structures of the steels used are represented by the
photographs in Fig. 28 where A and B are quite dendritic, C and D
nondendritic. The holes in the dies as shown in the photographs were
drilled out and not drifted, to eliminate any possible error that
might be caused by the various degrees of porosity in the center of
the die blanks.

Table X

Relative Wearing Properties of Dendritic and Nondendritic Steels Measured in Number
of Feet of Admiralty Condenser Tubing Drawn per 0.001 Inch Wear in
the Diameter of the Die.   Fig. 28

Die No.	C	Mn	Si	P	S	Cr	Ft. of tubing per 0.001 Inch Wear
A	1.03	0.32	0.18	0.017	0.012	0.14	675
B	1.02	0.36	0.34	0.010	0.009	....	1087
C	1.24	0.28	0.27	0.019	0.023	0.06	872
D	0.84	0.25	0.24	0.009	0.010	0.04	869
Average Dendritic dies							881
Nondendritic dies							871

The average of the three tests on each of the four blanks shown
in Fig. 28 are given in Table X.

It is to be noted that there is not much difference in the wearing

properties of dendritic and nondendritic steel when tested in this manner.

Summary and Conclusions

1. All steels under the commercial methods of manufacturing when cast are inherently dendritic to greater or less extent. The degree of dendritic condition appears to depend generally on the composition, the casting temperature, the degree of quietness with which metal solidifies, the viscosity and particularly the rate of cooling of the molten metal in the mold.

2. Structurally a dendrite is a complete crystal when formed under favorable conditions and a structural unit resulting from an attempt to form a complete crystal when cooled under unfavorable conditions. In either case it is built up of smaller crystalline units of orderly arrangement. A dendrite, therefore, arises from a preferred orientation of the smaller crystals of which it is made up and not from a selective freezing of the different elements in the metal. Any segregation follows the crystallographic habit of the dendrites but it does not give birth to the dendrites.

3. Chemically, a dendritic structure may or may not signify a segregation of various elements. For instance, a coring effect, that is, a chemical variation from the center of a grain towards the boundaries, is independent of the existence of visible dendrites and can happen in a negligibly dendritic as well as highly dendritic steel. Compare Figs. 20A and 20B with Figs. 21A and 21B.

4. After the formation of dendrites important changes take place in their inner structure, which although leave the outward form of the dendrites unaltered, they nevertheless impart improved properties to the metal. These changes occur during the process of granulation and secondary crystallization.

5. Mechanical properties of dendritic steel may or may not be inferior to those of nondendritic steel depending primarily on the relative amount of thermal and mechanical treatments which the steels have received and their respective microstructure. For instance it was noted that in the case of 1.00 per cent carbon steel dendritic steel showed lower physical properties than the same type but nondendritic steel in cast and normalized condition, but after forging them to 4 to 1 reduction and subjecting them to the same heat treatment, both dendritic and nondendritic steels showed the same microstructure and same physical properties although the dendrites still existed. The

same relation seems to hold between the microstructure and physical properties of dendritic and nondendritic low carbon chromium-nickel steel discussed in the paper. Therefore, microstructure and not the visible dendritic structure is the dominant factor in the development of physical properties.

6. In fatigue resisting properties as tested on a Stanton repeated impact testing machine 1.00 per cent carbon dendritic steel after identical thermal and mechanical treatment showed 16 per cent less fatigue resisting properties than the nondendritic steel.

7. In directional properties dendritic low carbon chromium-nickel steel showed greater difference than the nondendritic steel in the reduction of area, other properties being practically the same. In other words, dendritic steel when tested transversely showed 23 per cent less reduction of area than when tested longitudinally, while the nondendritic steel when tested transversely showed 9 per cent less reduction of area than when tested longitudinally.

8. In wearing properties dendritic and nondendritic carbon steels when tested for drawing dies showed practically no difference.

The authur wishes to acknowledge the courtesy of Col. T. C. Dickson, Commanding Officer of Watertown Arsenal for the results of Stanton repeated impact test given in Table IX.

DISCUSSION

Written Discussion: By V. O. Homerberg, associate professor of physical metallurgy, Massachusetts Institute of Technology, Cambridge, Mass.

Mr. Keshian has presented an interesting contribution to this important subject, and I wish to compliment the author on the excellence of his macrographs. Since he is a user rather than a manufacturer of steel his comments to the effect that the presence of a visual dendritic structure, as revealed by the hot acid etch, does not necessarily indicate that the material is defective and should be rejected, should dissipate to a considerable extent this prevalent idea. His experience that many tools showing a dendritic structure have given excellent service is in agreement with my observations.

The terms "dendrite" and "dendritic segregation" are often confused by many metallurgists. Dendrites may occur in both pure metals and in solid solutions, whereas, dendritic segregation may occur only in solid solutions. The presence of dendritic segregation is revealed by differential etching. Although I have noted dendrites in copper, purported to be 99.99+ per cent pure, the fact that they were visible after etching may be attributed to minute quantities of impurities or to slight amounts of gases.

In the author's discussion of dendritic and nondendritic steels it is apparent that he uses these terms in interpreting the appearance of the surface of steel sections after deep acid etching. Only the instantaneous solidification from the molten state, or the employment of some means to cause complete diffusion, if this is possible, of the segregated elements in the solidified metal can be considered as methods to eliminate dendritic segregation.

The results of his tests in the study of the directional properties are of interest. I would suggest that the results of impact tests be included in future investigations of this nature. I have conducted a number of tests in a study of the correlation of the impact properties with the directional properties of a number of steels. In some instances the differences in the impact values obtained when the stress was applied perpendicular to the direction of the "fiber" and when the stress was applied parallel with the "fiber" were about the same, whereas, in other instances marked differences were noted.

It is obvious from these observations that the size of the original dendrites and their heterogeneity are important factors in determining the final properties of the material since the specimens in each case received exactly the same heat treatment.

Written Discussion: By G. F. Comstock, Niagara Falls, N. Y.

Mr. Keshian's discussion of the crystalline and chemical nature of dendrites does not seem to help much in clarifying or explaining the phenomenon of their occurrence in steel. The arguments are hard to follow and sometimes contradictory. For instance, on page 323, line 2, it is stated that "all steels are by no means dendritic," while on page 333, line 5, we read that "the formation of dendrites of microscopic size can never be prevented." If this is true, it would seem that all steels must be dendritic to at least a certain degree. The author does not seem to keep clearly in mind throughout that what is called a dendritic structure is generally made evident on a large scale by a particular kind of etching, while a different kind of etching is used for showing the microstructure. This is a fundamentally important fact and must not be overlooked, as seems to have happened in Fig. 14 and its discussion.

It might well be asked what was wrong with the old conception of Stead, Sauveur, and other authorities that the dendritic structure is simply due to differences in solubility in the etching reagent, caused most generally by nonuniform distribution of impurities in solid solution in the steel? The author tries to attack this theory, but without substituting anything better. His quotation from the work of Stead on page 347 is misleading, and that from the work of Belaiew is incorrect. Stead proved many times that phosphorus might be a cause of dendritic structure; the statement he denied was that differences in solubility resulting in the appearance of dendrites were invariably due to phosphorus. He taught repeatedly that uneven distribution of phosphorus, nickel, chromium and other elements in solid solution in the steel led to differences in solubility; he also showed that differences in crystal orientation might have the same effect.

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The statement of Belaiew's, to which the author refers on page 347, is that "dendrites were revealed macroscopically in steels with very small amounts of phosphorus, from traces only in some cases, to the maximum amount of only 0.028 per cent." This is entirely different from the author's statement that Belaiew "found dendritic structure in steels free from any traces of phosphorus."

It is astonishing how often in the discussion of dendritic or streaky structure in steels the argument is advanced that because the phosphorus content of the steel is low as shown by chemical analysis of an average sample, the structure could not be due to phosphorus. Such an argument is entirely without weight, for the total amount of phosphorus in the steel has nothing to do with the question; it is simply the uneven distribution of the impurity that causes the difference in the severity of acid attack. Another point apparently missed by the author is that the impurity, in order to be effective as a producer of dendritic structure, must be in solid solution in the steel. The absence of this structure in the Armco iron illustrated by Fig. 16 is of no value whatever as evidence against a chemical cause for dendrites, because the only impurities shown to vary here are carbon and sulphur, both being not in solid solution in the iron.

The last part of the paper, dealing with the mechanical properties of dendritic steel, should be of considerable value, and the author has performed a useful service in publishing these data. The data on directional properties on page 365 seem especially pertinent, but their interpretation is open to criticism. The difference of about 10 per cent in elongation in the dendritic steel in favor of the longitudinal tests as compared with the transverse should not be overlooked. The conclusion that in all properties except reduction of area the dendritic steel tested transversely and longitudinally was as nearly alike as the nondendritic steel is not justified by the facts. In strength and "elastic limit" this is correct, but not in ductility as shown by both elongation and reduction of area. The author's results are actually more in agreement with the results in previous tests of the directional properties of streaky steel than his conclusions would indicate.

DR. ALBERT SAUVEUR:[1] Although I must disagree with Mr. Keshian's paper on some fundamental features of the subject covered, I appreciate the value of his work, and I desire to extend to him my congratulations.

One must take exception to the author's frequent references to nondendritic steel on the ground that such steel does not exist. All steels, regardless of composition, casting temperature or rate of cooling, solidify as a mass of dendritic crystals. Since these dendrites represent the solidification of a solid solution, they are necessarily heterogeneous when first formed, the axes containing less of some of the dissolved elements than the fillings. This chemical heterogeneity of the dendrite is generally referred to as dendritic segregation. While the well known phenomenon of diffusion tends to obliterate this lack of chemical homogeneity, it rarely effaces it altogether and we may then speak of persistent dendritic segregation. The dendrites themselves are destroyed as

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[1]Member A. S. S. T., Gordon McKay Professor of Metallography and Metallurgy at Harvard University, Cambridge, Mass.

independent crystalline units by the process of granulation taking place below
solidification. The existence of persistent dendritic segregation, however, makes
it possible to bring out the dendritic structure as it must have existed after
solidification through the selective action of suitable reagents which affect the
axes and fillings differently. When the author speaks of nondendritic steel he
must refer to those in which he was unable by the usual etching methods to
reveal a dendritic structure, an evidence not of absence of dendrites when the
steel solidified, but an evidence of the absence of persistent dendritic segregation.

The author's contention, that the belief that no selective etching can take
place without chemical variations, is not true, demands some clarifying. He is
concerned with dendrites and I believe it to be undisputable that if dendrites
were not chemically heterogeneous their internal structure could not be re-
vealed by any etching method. His failure to obtain dendritic structures in
some steels is a proof of the correctness of that assertion. I made an attempt
some years ago to explain the mechanism of the formation of dendrites and of
dendritic segregation,[2] in which it was shown that phosphorus was prone to
produce persistent dendritic segregation. The introduction of 0.170 per cent
phosphorus in steel containing 0.39 per cent carbon resulted in dendrites, the
fillings of which were ferrite and the axes pearlite, confirming Stead's assertion
of some twenty years ago that phosphorus (present in the fillings) drives the
carbon into the axes. The author's contention that phosphorus plays an unim-
portant part in causing dendritic segregation is, therefore, completely opposed to
the results obtained by us, and I believe by others. Stead's claim was not
that phosphorus causes the formation of dendrites, but that when present in
dendrites it promotes segregation. The author asks how the dendritic struc-
ture of pure metals can be explained. The answer is that the formation of
dendrites does not require the presence of impurities, but dendritic segregation
does. Because of its absence in pure metals, a dendritic pattern cannot be ob-
tained on etching. If it can, one may be sure that the metal is not pure. In-
deed, the method may be employed to test the purity of cast metals. In regard
to the significance of a banded structure in hot-worked metals, it would indeed
be unwarranted to claim that all parts exhibiting it are defective. Were it so,
it is quite probable that more than 90 per cent of all forged or rolled steel
articles would have to be so classified. The case should be stated differently.
The more pronounced the dendritic segregation, the more marked the banding
after hot work and the more marked the banding the greater the directional
properties. Other things being equal, the steel showing the least banding is to
be preferred, especially if the manufactured article is to be subjected to trans-
verse stresses.

O. W. Ellis:[3] There are four points which I would like to take up in
order. In the first place, I would like to call attention to the diagram shown
on page 340, which has been reproduced on a number of occasions in other papers
and in other places. I have with me what I had no intention of showing in
connection with this discussion, and that is a slide showing various equilibrium

[2]Dr. Albert Sauveur, "Crystallization of Iron and Its Alloys," TRANSACTIONS, American
Society for Steel Treating, Vol. 4, July 1923, p. 12.

[3]Member A. S. S. T., director of research, Ontario Research Foundation, Toronto,
Ontario, Canada.

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diagrams for the iron-carbon system. In this the thin lines represent approximately the positions of the solidus and liquidus as determined by Carpenter and Keeling after some slight alterations have been made in their position to bring them into line with what we know is the melting point of lines at the present day. The heavy line represents the solidus as determined by Gutowsky, the dotted line, a suggested solidus that the speaker is going to discuss more fully in connection with Mr. Jominy's paper.[4] It is of interest that both Mr. Jominy and the speaker are in agreement that this line represents very closely the true solidus for this system; i. e., the Gutowsky solidus is incorrect.

The point I wish to bring out in connection with this diagram is the peritectic reaction at 1486 degrees Cent. We all know that peritectic reactions are more powerful in their effect on dendritism than are the conditions existing during even the rapid cooling of solid solutions. The fact that a peritectic reaction occurs in many of these alloys is one, therefore, that should not be overlooked.

Another point is on page 345, where the author states "that a fairly well developed dendritic structure in steel can be detected by the unaided eye by merely smooth-machining the metal without further polishing or resorting to etching, showing a structural difference and not necessarily a chemical difference. "I am quite unable to see why chemical differences should not bring out apparent structural differences. Surely if we have dendrites having a low carbon content in the center and a high carbon content toward the outside, then machining should bring out the differences in chemical composition, simply because differences in chemical composition imply differences in physical properties (hardness, etc.). Those of you who have read the work of Carpenter and Elam on Admiralty Bronze will remember that by machining very carefully the surface of ingots of bronze, they were able to bring out the microstructure quite clearly without the necessity of etching at all. Variations in the types of atoms, or in the concentration of similar atoms within individual crystals, will surely be brought out by careful machining or polishing alone.

Another point—on page 351 Mr. Keshian states: "Furthermore, slow cooling which promotes dendritic structure, allows more time to the impurities to agglomerate and rise to the surface or the discard end of the ingot." If I judge rightly the work of Giolitti and the work of Sauveur, the very opposite is true. I quite agree with Mr. Keshian that rapid cooling will start crystallization at a greater number of centers than slow cooling, but I cannot see that rapid cooling in any way lessens the possibility of dendritic segregation. In fact, it tends to increase that possibility. The slower one cools a solid solution, the more uniform will be the structure of the individual crystals. In brief, rapid cooling tends to produce a large number of small crystals. These crystals, if and when dendritic segregation occurs, will be highly "segregated." On the other hand, slow cooling tends to produce large crystals but will at the same time lessen dendritic segregation in the individual crystals.

This is a subject worthy of much further investigation, one in which we are all concerned, and one which we should thank the author for having brought before the Society again in this interesting way.

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[4]See TRANSACTIONS, American Society for Steel Treating, Vol. 16, December 1929, p. 933.

F. G. FRISBIE:[5] I would like to make just a few remarks on this interesting paper. I have studied several years under Dr. Krivobok, who is intensely interested in dendritic segregation by artificial means, and also to find out how we could keep on producing it by artificial means.

Our method was to pour duplicate test blocks from a heat at a high temperature, about 3000 degrees Fahr. One of these blocks would be shaken out of the sand mold immediately or as soon as the metal was solidified, the other one was allowed to cool to black in the mold. Toward the end of the pour, when the metal temperature was down to about 2800 degrees Fahr. the same thing was repeated.

We found that the metal poured at a high temperature showed more evidence of dendritic segregation than the metal poured at a lower temperature.

We have taken these blocks and cut them into test bars and given them various heat treatments and, possibly contrary to general opinion, we have found that the physical values of the test bars annealed from the material that was greater in dendritic segregation have had higher physical values than those that were lower in dendritic segregation.

DR. ANSON HAYES:[6] May I raise a question in regard to the compositions of ingot iron as shown on page 348? I wish to state that the yearly average for composition of ingot iron is carbon 0.018, manganese 0.02, silicon a trace less than 0.005, sulphur 0.025. This analysis should be compared to the data given in connection with the figure on page 348. We have noted a variation which would put the composition within the ranges mentioned in this figure.

O. W. McMULLAN:[7] The remarks I want to make are about dendritic structure in regard to the rate of cooling. I know from examining a large number of phosphor-bronze worm gears which have been chilled in a metal mold that we find long dendrites in the chilled surface, much longer than in the interior, and also in cast iron we find a similar thing. I believe we also get a so-called columnar structure on cast steel. The explanation for this is that the outer surface cooling first becomes solid, and the interior remains liquid so that the crystals build up in one direction. There is nothing to interfere with the growth in the liquid metal so that they become much elongated in one direction; interference in other directions coming from crystals formed at the same time. These crystals are usually long and narrow, while the crystals in the interior may have almost as great or perhaps greater volume, but they are more nearly equiaxed and appear smaller. In regard to some of the tensile properties on dendritic and nondendritic steels I noticed the Brinell hardness was not given. That is a desirable thing to obtain when the tensile strength test is made, although the difference shows up in elongation and in reduction of area.

Author's Closure

I wish to thank the gentlemen who took part in the discussion of the paper. It is significant that the substance of the discussions and the majority of arguments pertain to the causes of dendritic structure and leave the results

[5]Member A. S. S. T., chemist, Duquesne Steel Foundry Co., Coraopolis, Pa.

[6]Member A. S. S. T., chief chemist, American Rolling Mill Co., Middletown, Ohio.

[7]Member A. S. S. T., chief metallurgist, Timken-Detroit Axle Co., Detroit.

of my experimental work regarding the effect of dendritic structure on the physical properties of steel unchallenged, in which, after all, lies our ultimate interest.

Mr. Comstock sees a contradiction between my two statements that "microscopic dendrites can never be prevented" that, "all steels are not dendritic." He, apparently overlooks the circumstances under which these statements were made. My first statement refers to steel as cast, whereas, the second statement refers to commercially finished product. When steel solidifies, under certain conditions, only microscopic dendrites are formed, which under subsequent thermal and mechanical treatments tend to break up and become obliterated in many cases. That is why all commercial steels are not dendritic.

Mr. Comstock reminds me of the importance of distinguishing between macro and micro etching. While I wish to thank him for this, I am unable to see any connection between this and his reference to my discussion of Fig. 14. That figure has been used to illustrate the well known granulation theory of Belaiew, where I have stated that the large single crystal of dendrite in Fig. 14A, according to our conception of the formation of dendrites, must have been originally a crystalline unit made up of smaller crystals of similar orientation; but due to the process of granulation taking place after solidification, we find it to be broken up into many grains of dissimilar orientation as viewed under the microscope and as shown in Fig. 14B.

Mr. Comstock questions the wisdom of criticising the old conception of the causes of dendritic structure in steel. Fortunately, science knows no orthodoxy and recognizes no dogma and so long as a theory, no matter how brilliantly conceived does not satisfactorily account for all the observed facts, only good can come from its discussion and criticism. It should have occurred to Mr. Comstock that the presence of phosphorus in steel as causing the dendritic structure has not been generally accepted but has been questioned by authorities such as Belaiew, Le Chatelier and others. Stead himself does not consider the presence of phosphorus as being necessary for the formation of dendrites.

I appreciate Mr. Comstock's calling my attention to the inaccuracy in my wording as not exactly expressing the statements of Belaiew and Stead in regard to phosphorus. In order to correct any misunderstanding that may have occurred in this connection, I quote these two gentlemen verbatim.

Stead says, "What was responsible for the difference in the solubility of different parts of polished steel? He (Dr. Stead) had been more or less made responsible for stating that it was invariably phosphorus. He had never made such statement. He had pointed out that nickel and chromium and other elements led to differences in solubility. It was not necessary to have any chemical variation in the metal, for pure iron of large crystalline structure showed differences of solubility in different crystals." *(Journal,* British Iron and Steel Institute, 1919, No. I, p. 292.)

Belaiew says, "In 1907 special experiments were undertaken by him to study the crystallization phenomena on pure carbon steel, that is, on steels as free from phosphorus, sulphur and oxygen as possible. The results of these

experiments were published in 1908 and 1909 in Russian, and subsequently were reproduced in various papers and lectures. These results all tend to show that dendritic crystallization was not only present in all cases, but could be revealed microscopically in all steels, . . ." And again he says, ". . . the dendritic character of crystallization in steels is of a most general occurrence, he (Colonel Belaiew) is of the opinion that that character is to traced neither to the particular influence of phosphorus, nor that of oxygen, nor any other impurity, but to the general character of primary solidification, as resulting from the iron-carbon equilibrium diagram." (TRANSACTIONS, American Society for Steel Treating, 1924, Vol. 5, pp. 219, 220.)

Mr. Comstock referring to the absence of dendritic structure in Armco iron, shown in Fig. 16, after reminding us that "the impurity, in order to be effective as a producer of dendritic structure must be in solid solution in the steel" he says, "The absence of this structure in the Armco iron illustrated by Fig. 16, is of no value whatever as evidence against a chemical cause for dendrites, because the only impurities shown to vary here are carbon and sulphur, both being not in solid solution in the iron." He obviously overlooks the fact that the amount of phosphorus, which goes into solid solution in the iron, is here more than what is considered necessary to produce dendritic structure. Furthermore, as a matter of fact carbon is found to go into solid solution with iron anywhere from 0.008 to 0.08 per cent according to some investigators.[8]

Mr. Comstock in regard to the relative transverse properties of dendritic steel says that "the difference of about 10 per cent in elongation in the dendritic steel in favor of longitudinal tests as compared with the transverse should not be overlooked"; but it is to be remembered that in the same table, that is, Table VIII, similar difference exists between the transverse properties of nondendritic steel, in which case the difference is 7 per cent in favor of longitudinal tests, thus making the entire difference only 3 per cent in favor of nondendritic steel which is well within the experimental error and justifies my conclusion that the transverse properties of dendritic and nondendritic steels under consideration are about the same except in the reduction of area.

I wish to thank particularly Dr. Homerberg and Prof. Sauveur for their valuable contributions. I am glad to know that Dr. Homerberg's observations agree with my own that many tools showing dendritic structure have given excellent service. The distinction which Dr. Homerberg makes between the term "dendrites" and "dendritic segregation" is of extreme importance and their proper use would certainly prevent much of the confusion from taking place in the discussion of this subject.

The term nondendritic in this paper has been used in the same meaning as Dr. Homerberg understands, which I have explained more in detail in my reply to Prof. Sauveur.

I thoroughly agree with Dr. Homerberg on the importance of the transverse impact test. The fact that some of his impact tests have shown wide differences in the transverse properties of "fiber" and some have shown about the same properties, is significant. This tends to confirm my conclusion that

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[8]T. D. Yensen, "Pure Iron and Allotropic Transformations," *Heat Treating and Forging*, 1929, March, p. 321.

the mechanical properties of dendritic steel may or may not be inferior to those of nondendritic steel, depending primarily on the amount of thermal and mechanical treatment which the steels have received and on their respective microstructure.

Prof. Sauveur takes exception to my using the term "nondendritic," saying that such steel does not exist. He, evidently, includes in this statement all steels in cast as well as in finished condition.

From Figs. 9 and 10 it will be seen that if the steel shown in Fig. 9 is dendritic, the one in Fig. 10, certainly, is not, although it is not altogether free from tiny, microscopic dendrites. The term nondendritic, however, has not been used as to signify the absolute absence of these invisible, microscopic dendrites, any more than many other technical terms, such as noncorrosive, nondeforming, nonchangeable, stainless, etc., signify the absolute absence of the properties to which they refer. If we were to use all the technical and even scientific terms in their absolute meaning very few of them would stand the test.

Furthermore, if Prof. Sauveur's objection refers to all the finished steels as well as to cast steels, the use of the term is even more justified, because if a steel, in the as-cast condition, has only microscopic dendrites, like the one shown in Fig. 10, and then, is subjected to thermal and mechanical treatment during the process of manufacture, these diminutive microscopic dendrities break up and become obliterated, then the finished steel becomes practically nondendritic.

The question which Prof. Sauveur raises regarding the interpretation of selective etching as an indication of chemical homogeneity is vital and I, for a moment, do not question the fact that the solidification of solid solutions necessarily implies chemical heterogeneity of dendrites. Nevertheless there are certain difficulties in the way to believe that the revealment of dendrites by selective etching does always and invariably mean, as Prof. Sauveur asserts, chemical heterogeneity even after they have been subjected to the homogenizing effects of diffusion and thermal and mechanical treatments. Prof. Sauveur's view that if the diffusion and other thermal and mechanical treatments had produced a homogeneous structure we would not be able to disclose dendrites by selective etching methods such as by cupric reagents, excludes entirely the idea as expressed by others that the difference of solubility might also be due to the difference of electro-potential between the two adjacent grains arising from their difference in orientation thus causing the copper to deposit more on certain parts than others.[9]

Again if the dendrites as revealed by selective etching invariably means chemical heterogeneity, the greater the dendritic condition the greater must be the heterogeneity. Consequently we should get a much greater difference between the transverse properties—whether tensile, or impact—of intensely dendritic steels and the steels that are nearly free from dendrites than we actually do, but we not only do not find such difference as shown by Table VIII and Fig. 22, but, on the contrary, we often get better physical properties

[9]Stead, *Journal*, Iron and Steel Institute, 1919, No. 1, p. 292. Arnold, *Journal*, Iron and Steel Institute, 1921, No. 1, p. 292. Howe, *Journal*, Iron and Steel Institute, 1921, No. 1, p. 295.

from dendritic steels than from less dendritic steels as mentioned by Mr. Frisbie, and at times we get about the same, as stated by Dr. Homerberg.

On the ever-discussed influence of phosphorus, Prof. Sauveur says, "The author's contention that phosphorus plays an unimportant part in causing dendritic segregation is, therefore, completely opposed to the results obtained by us, and I believe, by others." This, undoubtedly conveys a different idea than that I have expressed in the paper. What I had said about phosphorus, pertains to the formation of dendrites and not to dendritic segregation, which terms, as I understand, and as Dr. Homerberg has stated in his discussion, have entirely different meaning, whereas, Prof. Sauveur uses them interchangeably, as evidenced from his following statements. Referring to his experiments, he says, "in which it was shown that phosphorus was prone to produce persistent dendritic segregation." Again he says, "the introduction of 0.17 per cent phosphorus in steel containing 0.39 per cent carbon resulted in dendrites." Then he continues saying that "Stead's claim was not that phosphorus causes the formation of dendrites, but when present in dendrites it promotes segregation." I have already quoted Belaiew's view that phosphorus is not to be looked as the cause of dendritic structure in steel. Therefore, if Prof. Sauveur holds the view that phosphorus gives rise to dendrites he opposes both Belaiew's and Stead's views, also the view I have expressed, which was, that (page 370) "Any segregation follows the crystallographic habit of the dendrites, but does not give birth to the dendrites."

That it cannot be the phosphorus that gives rise to dendrites (not dendritic segregation) can also be induced from a comparison of steel shown in Fig. 15, and the Armco iron shown in Fig. 16, with the steel shown in Fig. 17. The first two metals have an amount of phosphorus that is higher than generally claimed to be sufficient to produce dendrites. Yet neither of these materials show the presence of dendrites. On the other hand the steel shown in Fig. 17, made by melting electrolytic iron with sugar charcoal in vacua, by Prof. Sauveur himself and Dr. Krivobok, show well developed dendrites, despite the fact that it must have been low in phosphorus and considerably lower than the amount of phosphorus in Figs. 15 and 16.

Prof. Sauveur's comments on the effect of dendritic structure upon the physical properties of the steel is significant as coming from an authority who has taught us so much on this subject, in which he says that it would be unwarranted to call defective any hot-worked metal on account of banded structure in it. This is gratifyingly less drastic than his previous attitude that "persistent dendritic segregation is always harmful and should, if possible, be prevented or reduced."[10]

Dr. Ellis brings out the question of accuracy of the iron-carbon diagram in Fig. 13. As this diagram is the one generally accepted and was here used to illustrate the granulation zone of Belaiew, its further discussion here would only have a negative interest to the subject under consideration, although I am in sympathy with his views regarding the necessity of revising the diagram.

In connection with my statement on the detection of dendritic structure by etching and by machining, Dr. Ellis says that he is not quite able to see why

[10]Dr. Albert Sauveur, TRANSACTIONS, American Society for Steel Treating, 1923, Vol. 4, p. 45.

chemical difference should not bring out apparent structural difference. Quite true, but we are concerned with the interpretation of the relief effect. The fact that we can bring out relief effect by etching does not necessarily mean that we should have chemical difference; because we can also bring out that relief effect without etching, such as by polishing or machining. In other words, chemical difference may cause structural difference; but structural difference is not always caused by chemical difference. That can also be caused by differences in the grouping or a difference in the orientation of the perfectly homogeneous metal. For that reason, selective etching is not an absolute proof of chemical heterogeneity of the metal.

In his reference to the effect of the rate of solidification, Dr. Ellis brings forth an unquestionably important phase of the subject. His remarks, however, implies to the phenomena of segregation as resulting from the solidification of solid solutions; but not to the separation of insoluble inclusions to which I have referred in my statement which he has quoted. Therefore, his reference, in that connection, to Giolitti and Sauveur, is erroneous.

In the matter of separation of inclusions, I believe, there can hardly be any question that these impurities have greater chance to agglomerate and rise to the surface by slow cooling than would be the case if the solidification was so rapid that the mushy state would be reached too quickly and thus the impurities would be prevented from coagulating and rising to the surface.

Giolitti, referring to the solidifying metal, says, "During these relatively quiet periods, all suspended slag globules tend to separate on account of a difference in density, a process analogous to that taking place when settling a turbid liquid. . . . These large drops ascend toward the surface of the denser supporting liquid, at a more rapid rate the larger their dimensions."[11] Rapid cooling would greatly reduce this self refining process of the molten metal.

Dr. Ellis, referring to the effect of the rate of solidification on the production of dendritic segregations, states the fact, correctly, from the standpoint of the solidification of solid solutions and at least in the case of some steels, when he says that, "rapid cooling tends to produce a large number of small crystals. These crystals, if and when dendritic segregation occurs, will be highly segregated. On the other hand slow cooling will produce large crystals, but will at the same time lessen dendritic segregation in the individual crystals." From this statement of Dr. Ellis, it would at once appear that slow cooling is better than rapid cooling, and from the practical point of view a highly dendritic steel would be more homogeneous than the steel with very small dendrites and there would be, perhaps, no more question of dendritic steel. But contrary to Dr. Ellis' claim neither Giolitti nor Sauveur recommend larger dendrites if I understand them from their following statements.

Sauveur says: "We should endeavor to produce in our finished castings dendrites as small as possible—."[12]

Giolitti, after pointing out the homogenizing effect of slow cooling, is more explicit in his preference for smaller dendrites. He says: "Thence it can

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[11]Giolitti, "Heat Treatment of Soft and Medium Steels," McGraw-Hill Book Company, 1921, p. 134.

[12]Dr. Albert Sauveur, "Crystallization of Iron and its Alloys," TRANSACTIONS, American Society for Steel Treating, 1923, Vol. 4, p. 45.

be deduced, all other conditions being equal in a determined crystalline mass, that the gain in homogeneity induced by a given 'reheating for diffusion' will be greater the larger has been the number of centers per unit of volume around which the crystalline elements have formed; in other words, the greater has been the frequency of nuclei in the molten mass under observation."[13]

Giolitti goes even further, and in section 82 and page 108 of his book just quoted, after referring to the process of quenching for normalizing, he says: "An analogous proceeding might also find industrial application as a means to regulate the process of primary crystallization, suppressing the effects of liquation or segregation which usually accompany solidification."

The opinion on the effect of rapid cooling on the homogeneity of steel as expressed in the recently published third report of the British Iron and Steel Institute Committee on the Heterogeneity of Steel Ingots is significant which reads as follows: "Two reasons for the greater homogeneity of nickel and nickel-chromium steels, as compared with carbon steels, arise out of this investigation. In the first place, the liquidus and solidus curves for the alloy steels, i. e., nickel and nickel-chromium, lie nearer together with respect to temperature than they do in the case of plain carbon steels. This necessarily means that solidification takes place over a narrower range of temperature, and therefore relatively more rapidly."[14] This would indicate that rapid rate of solidification is desirable.

Mr. Frisbie's contribution is valuable and interesting, and I am glad to know that his results confirm my own. The fact that he found the steels of greater dendritic structure showed greater physical properties than the steels less dendritic confirms my results with the chromium-nickel steel of S.A.E. 3312 type reported in the paper.

Dr. Hayes finds the analysis of Armco iron shown in Fig. 16 not in agreement with his own. Since the presentation of the paper a double check by two different reliable sources showed that their average were not as low as Dr. Hayes considers should be. The explanation, I believe, lies in the segregated condition of this particular piece of Armco iron as clearly shown by the etched section of the specimen in Fig. 16. It would appear that this piece of metal must have come from near the segregated or discard end of the ingot, and, unless Dr. Hayes makes his analysis on every ingot and just at or near that particular section of the segregated end, the heat or ladle analysis would only indicate the average composition of the ingot, but would not tell him the composition of the segregated section which quite naturally would be higher than the highest heat analysis.

Mr. McMullan mentions the desirability of giving the Brinell hardness of specimens in connection with physical tests. This refers to the slides shown during the reading of the paper. The importance of this fact was always born in mind during the tests, and although the Brinell numbers were not shown in the slides but were given in the original paper.

[13]Giolitti, "Heat Treatment of Soft and Medium Steel." McGraw-Hill Book Co., 1921, p. 27.

[14]Third Report on the Heterogeneity of Steel Ingots. *Journal*, Iron and Steel Institute, 1929, No. 1, p. 337.

GRAPHITIZATION OF PREQUENCHED WHITE CAST IRON

By H. A. Schwartz, H. H. Johnson, and C. H. Junge

Abstract

The paper constitutes a study of the acceleration of graphitization produced by the prequenching of the hard iron before annealing, as observed by the present authors and others. The study is largely on the basis of observations of the progress of graphitization with time and leads to the conclusion that the reactions involved are interfacial in character and represent the rate of solution of cementite in the metallic matrix. It is shown that the diffusivity of carbon in iron is not necessarily altered by the prequenching operation, but that the migration of carbon is much accelerated, due to a greatly increased number of graphite nuclei, with the accompanying decrease in migratory distance. Data are given to show that there is no ground for assuming that prequenching alone increases the carbon content of the phase which appears to be cementite, but that, on the contrary, there is some ground for the belief that between A_1 and the solidus the phase in equilibrium with austenite is not the pure iron carbide, but, perhaps, a solid solution of iron in iron carbide, so that the concentration of carbon in the resulting phase is less than $6\frac{2}{3}$ per cent.

Announcement is made of a prospective further investigation of the latter point by another observer.

Introduction

ONE of us has studied for some years the effect of preliminary heat treatments applied to white cast iron upon the subsequent graphitizing rate. Publication was withheld pending the result of a patent application[1] on the process and in the meanwhile Saito and

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[1] H. A. Schwartz, U. S. Patent 1,688,438, October 23, 1928.

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Sawamura[2] have published the results of their own experimentation on very similar lines, making superfluous a publication on our part of work which almost precisely duplicates theirs.

It is the purpose of the present paper to gather together certain material collected in our study of this problem dealing with subjects not touched upon by the Japanese investigators and calculated to add somewhat to our knowledge of the mechanism of graphitization in general and to the process of graphitization under these special conditions in particular. Two earlier articles from this laboratory[3,4] may be consulted for an exposition of the relations so far developed between the form of the time-graphite curve and the reactions underlying graphitization.

The subject matter of the present paper will fall naturally into two subdivisions, (1) a study of the effect of prequenching upon the carbides existing in white cast iron and (2) a study of the rate of formation of graphite in prequenched white cast iron from the view point of attempting to recognize the character of the reactions going on.

EFFECT OF PREQUENCHING UPON COMPOSITION OF THE CARBIDES

In an earlier paper[5] it was recognized that the carbides isolated from quenched white cast iron by an electrolytic method there described were much richer in agraphitic carbon than the theoretical $6\frac{2}{3}$ per cent corresponding to the composition of cementite. A repetition of similar experiments on material quenched from 1650 degrees Fahr. (900 degrees Cent.) or above confirmed these earlier results. Carbides were obtained of very erratic composition and very high concentrations of "combined carbon". The value of $14\frac{1}{2}$ per cent for combined carbon was, perhaps, more frequently encountered than any other, but some higher and many lower values were obtained. It was later found that prequenching was not an essential in obtaining these preparations, but that they were sometimes

[2]Saito and Sawamura, "A New Phenomenon Concerning the Graphitization of White Cast Iron and Its Application to the Manufacture of Black Heart Malleable Castings," reprinted from the Memoirs of the College of Engineering, Kyoto Imperial University, Vol. V, No. 1, 1927, page 2.

[3]H. A. Schwartz, "Graphitization at Constant Temperature," TRANSACTIONS, American Society for Steel Treating, Vol. 9, No. 6, June, 1926, page 883.
[4]Schwartz and Johnson, "Graphitization at Constant Temperature Below the Critical Point," TRANSACTIONS, American Society for Steel Treating, Vol. 10, No. 6, December, 1926, page 965.

[5]H. A. Schwartz and A. N. Hird, "Chemical Equilibria During Solidification and Cooling of White Cast Iron," Transactions, American Institute of Mining and Metallurgical Engineers, Vol. LXXI, 1925, page 470.

obtainable from "as cast" metal which had cooled in the mold at no unusual rate. Through the courtesy of J. H. Critchett of the Union Carbide and Carbon Laboratories, Inc., Long Island City, New York, we had an opportunity to avail ourselves of the kind co-operation of E. C. Bain, then of that laboratory, in an X-ray spectrometric study of these carbides as compared with a normal cementite prepared in this laboratory from unquenched white iron. Our sincere thanks are due to Messrs. Bain and Critchett for this most helpful co-operation. In the light of the X-ray spectograms obtained it proved impossible to show any difference in space lattice or parameter between cementite and these higher carbon derivatives. The erratic character of the results, together with the metallographic evidence to be presented shortly, motivated a further study of the electrolytic decomposition of quenched white irons.

We have been so fortunate as to enlist the interest of Dr. John White of the Rose Polytechnic Institute in the problem of what takes place during the electrolytic decomposition of cementite and it is hoped that experimental work under his direction will be productive of further knowledge of the matter.

In the meanwhile we will content ourselves with the inclusion of Table I, showing the results of various electrolytic conditions on a given hard iron and a few words of comment.

Table I
Results of Various Electrolytic Conditions on a Given Hard Iron

Current Density 0.00115 amp/cm^2

	As Cast	Prequenched from 1697 Degrees Fahr. (925 Degrees Cent.)
Voltage, max.	0.43	0.47
Voltage, min.	0.15	0.19
Voltage, mean	0.34	0.38
Total carbon	5.80	6.50
Graphite	0.06	0.22
Agraphitic carbon	5.74	6.28
Fe	85.93	84.26
Si	1.41	2.16
Mn	0.28	0.32
S	0.06	0.06
Mn in excess of MnS	0.18	0.22
Oxygen by Difference	6.52	6.70
Silicon volatile in Cl	0.31	1.00
Si as SiO$_2$	1.10	1.16

Current Density 0.00317 amp/cm^2

	As Cast	
Voltage, max.	0.82	0.77
Voltage, min.	0.32	0.28
Voltage, mean	0.60	0.52

Total carbon	6.90	6.48
Graphite	0.07	0.23
Agraphitic carbon	6.83	6.25
Fe	84.54	80.92
Si	1.54	2.15
Mn	0.36	0.31
S	0.05	0.05
Mn in excess of MnS	0.27	0.22
Oxygen by Difference	6.61	10.09
Silicon volatile in Cl	0.83	0.41
Si as SiO_2	0.71	1.74

Current Density 0.00533 amp/cm²

Voltage, max.	0.76	0.87
Voltage, min.	0.30	0.38
Voltage, mean	0.61	0.62
Total carbon	12.25	9.05
Graphite	0.26	0.21
Agraphitic carbon	11.99	8.84
Fe	68.41	76.20
Si	3.55	2.96
Mn	0.30	0.33
S	0.06	0.08
Oxygen by Difference	15.43	11.34
Silicon volatile in Cl	1.60	1.28
Mn in excess of MnS	0.20	0.19
Si as SiO_2	1.95	1.68

Current Density 0.00764 amp/cm²

Voltage, max.	1.09	1.06
Voltage, min.	0.40	0.41
Voltage, mean	0.76	0.72
Total carbon	13.52	10.37
Graphite	0.08	0.36
Agraphitic carbon	13.44	10.01
Fe	66.74	75.36
Si	3.97	3.07
Mn	0.30	0.24
S	0.04	0.04
Mn in excess of MnS	0.23	0.17
Oxygen by Difference	15.43	10.92
Silicon volatile in Cl	1.75	2.14
Si as SiO_2	2.22	0.93

Current Density 0.00072 amp/cm²

	As Cast	As Cast
Voltage, max.	0.38	10.0
Voltage, min.	0.18	8.5
Voltage, mean	0.33	8.9
Total carbon	5.88	6.02
Graphite	0.05	0.06
Agraphitic carbon	5.83	5.96
Fe	86.49	86.21
Si	1.40	1.27
Mn	0.34	0.13
S	0.05	0.05
Mn in excess of MnS	0.25	0.04
Oxygen by Difference	5.84	6.32
Silicon volatile in Cl	0.45	0.39
Si as SiO_2	0.95	0.88

Be it noted that all the samples contain very large amounts of oxygen, also that their silicon content, as FeSi, is somewhat constant and of the order of magnitude of 1.00 per cent.

The ratio of the Si of SiO_2 to the total O by difference is always of the same order of magnitude, about 0.143, suggesting a sharing of O between Fe and Si in an approximately constant proportion.

The carbon, silicon, and oxygen concentrations increase with current density, but not with potential. There arises some question whether cementite can be prepared by this method without gross contamination by oxygen or whether, perhaps, this contamination occurs only in the presence of silicon and represents the decomposition product, by hydrolysis, of a silicon compound of iron.

We are now inclined to think that the "percarbides", if we may use that phrase, which we isolated from quenched white iron, do not exist as such in the metal before electrolysis, but are the results of electrolytic attack of the carbide present in the specimens being investigated. This conclusion is very largely based upon the following metallographic examination. White cast iron specimens containing about 0.35 per cent silicon and about 0.30 per cent manganese and of varying carbon concentration were prepared by melting together in a small arc furnace the appropriate amounts of wash metal, Armco iron, ferrosilicon and ferromanganese. The silicon was introduced to permit of making reasonably sound specimens and the manganese to take care of the sulphur content of the constituents of the charge. The specimens were in the form of the A. S. T. M. tensile specimen for malleable iron[6] and were quenched in water from 1795 degrees Fahr. (980 degrees Cent.). After quenching the specimens were etched in boiling alkaline picrate, in order to bring out the boundaries of the contained cementite masses. It is worthy of note that the carbide of quenched samples is not nearly as readily attacked by this reagent as that of as cast samples.

In Figs. 1 and 2 are shown two specimens of the same iron etched together in alkaline picrate, the former being in the as cast condition and the latter in the quenched state. The photomicrographs are presented in demonstration of the fact that the quenched carbides apparently differ in chemical behavior from normal cementite.

By the method of intercepts, traversing a total of 160 inches on photomicrographs at approximately 300 diameters, we determined

[6]"Standard Specifications for Malleable Castings A 47-27," American Society for Testing Materials, A. S. T. M. Standards, Part I, Metals, 1927, page 438.
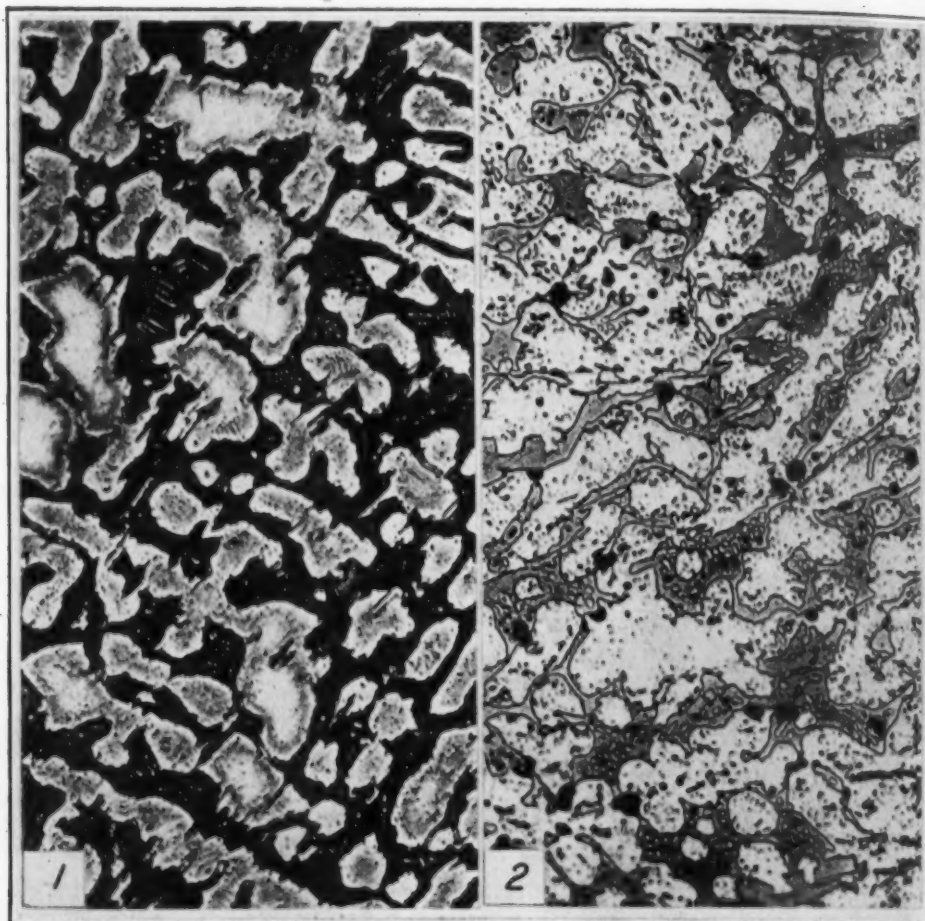
Figs. 1 and 2—Photomicrographs of Two Specimens of the Same Iron, Etched Together in Alkaline Sodium Picrate. Fig. 1 As Cast. Fig. 2 Quenched.

the volume of so-called cementite, in each of the specimens with the results shown in Table II, together with determinations of density by the Archimedian method on both the quenched and unquenched metal.

It seemed to follow from the phase rule that the carbide in all the samples should be of identical composition and density and it is possible, from this postulate, to compute these constants. The carbon content of the aggregate is a straight line function of the carbide content, by volume, as is also the density of the aggregate. Consequently, the carbon content is also a straight line function of the density. Evaluation of the tabular values from Table II by the method of least squares gives as the most probable relations

$$V_c = 28.537 \ (C - 1.766)$$

and

$$\rho = 8.017 - .0959 \ C$$

Table II

Density of Cementite Determined by Archimedian Method

| Total Carbon | Hard Iron Analysis (After Prequenching) | | Before Prequenching | Not Corrected for Graphite | After Prequenching Corrected for Graphite | Volume Carbide |
	Graphitic Carbon	Agraphitic Carbon				
3.12	0.07	3.05	7.7224 ⎫	7.7145 ⎫	7.7264 ⎫	...
3.12	0.07	3.05	7.7170 ⎬	7.7224 ⎬	7.7344 ⎬	40.0
3.01	0.08	2.93	7.7272	7.7217	7.7354	34.7
2.74	0.07	2.67	7.7388	7.7641	7.7763	26.9
2.31	0.08	2.23	7.7301	7.7902	7.8041	16.4
2.13	0.05	2.08	7.7524	7.8137	7.8225	10.1

where V_c is the percentage (by volume) of carbide in the aggregate, ρ the density of the aggregate and C its carbon content, correction having been made for graphite present.

Evidently V_c becomes zero when $C = 1.766$, so that this carbon concentration is that of the saturated austenite at the quenching temperature. This value is higher than the accepted locus of A_{cm} would indicate, the latter suggesting a value near 1.5 per cent. When V_c is 100 carbon becomes 5.27, which is then the probable carbon concentration in the carbide as compared with the expected value of 6.67 for cementite.

The values of ρ for 1.766 C and 5.27 carbon respectively are 7.848 and 7.512, corresponding fairly well with the probable densities of martensite and cementite.

We do not remember to have seen any formal demonstration that the carbide phase in eutectiferous white cast irons at temperatures between A_1 and the solidus is necessarily pure iron carbide, nor are we aware of any determinations of the A_{cm} line except in steels having a carbon content between the eutectoid composition and the maximum solubility of carbon in austenite. We do not regard the low carbon value obtained for the carbide as especially disconcerting feeling that it may be an expression of a limited mutual solubility of cementite and gamma iron.

It is suggested that there may be need of a new line in the metastable diagram, having its lower end at carbon 6⅔ per cent at and below the A_1 point and curving to the left as it rises to the eutectic melting point. The solid phase to the right of this line we construe to be a solid solution of gamma iron in cementite. Alternatively the carbide stable between A_1 and the eutectic freezing point might be Fe_4C or Fe_7C_2. In order that our findings may be verified A. Allan Bates, instructor in metallurgy at Case School of

Applied Science, has kindly consented to study this matter in greater detail and it is hoped that his work will result in further information with regard to the locus of this new line if, indeed, it shall be found to exist as well as in a check upon our present understanding of the location of the A_{cm} line. These matters must remain a subject of possible future publication by Mr. Bates upon the completion of his work.

If our present conception of the character of the carbide phase in the quenched sample should be confirmed, then any unusual
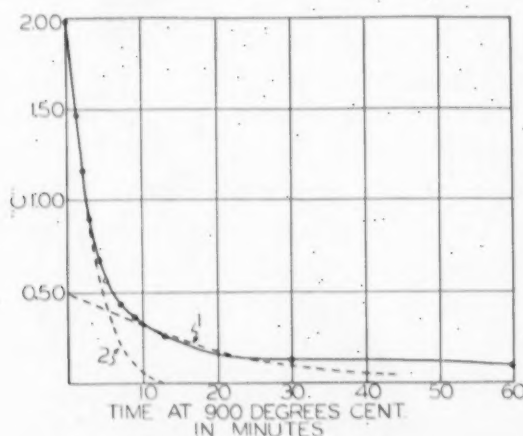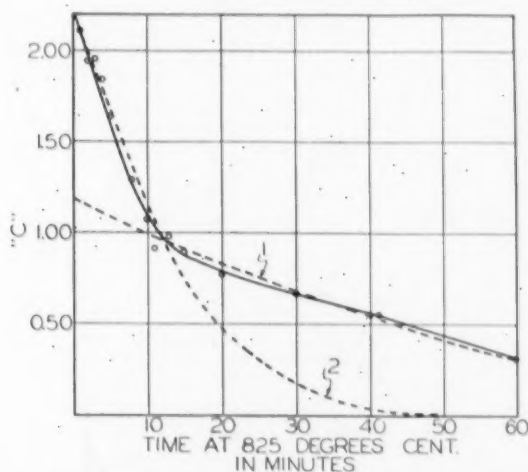

Fig. 3—Curves Showing Relation Between Graphitizing Temperature and Graphite Formed. Curve 1, $C = 0.4967 - 0.01998t + 0.000221t^2$. Curve 2, $C = (1.242 - 0.09327t)^3$.

"percarbides" obtained by electrolysis may represent the effect of dissolved gamma iron upon its electrolytic composition.

PROGRESS OF GRAPHITIZATION IN QUENCHED WHITE CAST IRON

For the relation between the time during which the sample is submitted to graphitizing temperature and the graphite formed a series of experiments were carried on, using as a raw material a white cast iron containing 2.95 per cent carbon and 1.17 per cent silicon, the manganese, sulphur, and phosphorus being such as would be common in ordinary malleable practice. The specimens were in the form of thin plates approximating a thickness of ⅛ inch (3 millimeters). The heat treatments were all conducted by immersing these thin specimens in a lead bath of such size that its thermal capacity would prevent any measurable change of temperature on immersing the specimens. Time was considered to begin at the moment of immersion. It is, of course, academically true that in

the earlier stages the temperature is rising from the room temperature of the specimen to the temperature of the lead bath. Under the circumstances this heating is very rapid and no attempt at correcting for this theoretical lack of precision has been deemed justified.

It was found if specimens were prequenched from 1650 degrees Fahr. (900 degrees Cent.) or more the subsequent rate of graphitization was so great as to occasion considerable difficulty in removing specimens at sufficiently short time intervals to follow the reaction

Fig. 4—Curves Showing Relation Between Graphitizing Temperature and Graphite Formed. Curve 1, $C = 1.1929 - 0.0206t + 0.0001006t^2$. Curve 2, $C = (1.3098 - 0.0261t)^3$.

velocity. Prequenching from 1560 degrees Fahr. (850 degrees Cent.) was accordingly resorted to and proved quite satisfactory. In Figs. 3, 4, 5, and 6 are shown graphically the results obtained. The time intervals are accurate to perhaps half seconds. The graphite determinations were made on pieces covering the cross section of the specimen, but it is unlikely that they are reliable especially for considerable amounts of graphite with a greater precision than some 0.04 per cent. In the figures graphitization is represented by the decrease in the available combined carbon as it progresses. This is the quantity represented by the symbol k in our earlier papers before this society, reference to which has already been made. The carbon available for graphitization is considered to be the total carbon of the specimens less the traces of graphite present as the result of the prequenching treatment and less the amount of carbon which is soluble in austenite at the temperature under consideration. In Figs. 3, 4, 5, and 6 certain curves are shown in

comparison and marked with the equations of which they are the loci. The significance of these curves will appear a little later in this discussion.

Considering it likely that the time to produce a given degree of graphitization would be proportional as between the different curves, irrespective of the specific degree of graphitization involved,
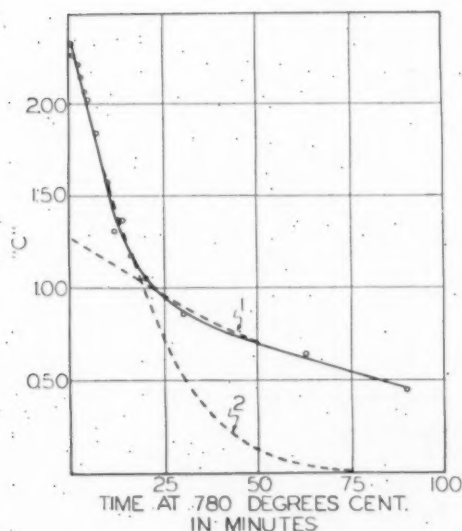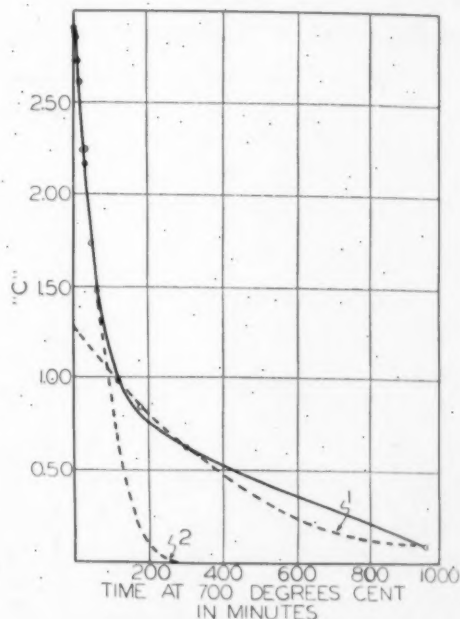

Fig. 5—Curves Showing Relation Between Graphitizing Temperature and Graphite Formed. Curve 1, C = 1.2697-0.0136t + 0.00005096t². Curve 2, C = (1.3382-0.01751t)³.

Fig. 6—Curves Showing Relation Between Graphitizing Temperature and Graphite Formed. Curve 1, C = 1.2776-0.00259t + 0.000001409t². Curve 2, C = (1.4379-0.00502t)³.

we attempted to superimpose the curves by the expedient described in our earlier reference.* The agreement was fair, but perhaps not quite so close as under the former conditions. Fig. 7 shows the result on semi-logarithmic co-ordinates. The visual average of the 4 curves has been drawn in for comparisons. By the method of least squares the equation of this average curve was found to be

$$(c + .00033) (t + 1.0125) = .06576 \qquad (1)$$

This equation is expressed in the same general units used in the reference.* Neglecting the very small numerical quantity in the first factor on the left side of the equation and noting that the numerical term of the second factor is very close to unity, it represents a rectangular hyperbola, having for its asymptotes the time axis and

*See foot note 3.

the line $t = -1$. Although this average equation may be a convenient means of approximating the relation between graphite and time, it is believed to be meaningless chemically.

THERMAL COEFFICIENT OF GRAPHITIZING RATE

In the absence of a demonstrated equation representing the locus of the graphite-time curve we may well derive our knowledge of the effect of temperature on graphitizing rate under the present
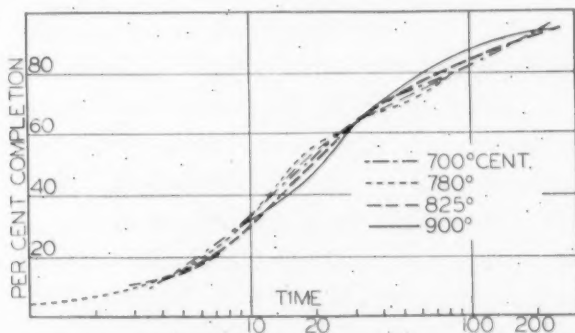

Fig. 7.—Graphitization Curves for Prequenched Hard Iron Superimposed.

conditions by observing how far we have had to transpose the curves in Fig. 7, in order to obtain the best possible agreement. The linear distance by which we have to transpose these curves represents, when measured against the logarithmic scale to which the curves are plotted, the number by which the time to attain a given degree of graphitization at one temperature must be multiplied to calculate the time required to obtain the same degree of graphitization at the temperature upon whose curve we wish to superimpose the first curve. The amount of transposition was determined as the average linear distance measured parallel to the time scale between the two curves to be superimposed, this distance being expressed by comparison with the logarithmic scale of the figure. The factor for the time intervals involved is shown in Table III.

DISCUSSION OF DATA

In deciding upon the interpretation to be placed on the graphitizing curves given above we may avail ourselves of certain principles. In the process of converting cementite into free carbon at least the following operations must be carried out. The carbon must be separated from its chemical union with iron, it must go into solution in some form at the boundary between the cementite and the austenite

Table III
Time Interval Factors

Temperature Range	Temperature Interval	Factor	Log_{10} Factor	Log_{10} Coefficient for 10°C
825°C ⎱ 780°C ⎰	45°C	1.171	0.06856	0.01521
900°C ⎱ 780°C ⎰	120°C	1.845	0.26600	0.02217
780°C ⎱ 700°C ⎰	80°C	1.372	0.13735	0.01717
			Average	0.01818
			Coefficient	1.043

or boydenite,[7] it must migrate from the location of a cementite particle to that of a carbon crystal and must separate from the solid solution at the boundary between solid solution and free carbon. In addition to these chemical processes it is also necessary that nuclei of free carbon should form either simultaneously or progressively upon which the carbon from the graphitizing reaction can crystallize.

In our earlier publication* it was shown that the time-temperature relations ordinarily encountered are those corresponding to the process of migration of the carbon through the solid solution. Therefore, since the rate of a sequence of reactions is of necessity determined by the rate of the slowest reaction involved, the conclusions of the earlier paper apply so long as migratory rate is the slowest of the steps involved in the process of graphitization. Under circumstances where the rate of graphitization is determined by the rate of decomposition of cementite, the relation of time to graphite formed would be logarithmic corresponding to the formula for the rate of a monomolecular reaction. Such a condition may, perhaps, be found in the attainment of equilibrium below A_1 of material which had previously attained equilibrium at a temperature above that point. Reasons have previously been given† why these observations may not be entirely conclusive, owing to the fact that equations for other types of reactions could be found so nearly identical that the logarithmic equation of the simple reaction as to be experimentally indistinguishable. The individual curves shown in Figs. 3, 4, 5, and 6 cannot be brought into harmony with the assumption that the

[7]Schwartz, Payne, Gorton and Austin, "Conditions of Stable Equilibrium in Iron-carbon Alloys," *Transactions,* American Institute of Mining and Metallurgical Engineers, Vol. LXVIII. 1923, page 925.

*See foot note 3.

†See foot note 4.

logarithm of the combined carbon remaining is a straight line function of the time, even though the curves bear some general resemblance to the equation of this type, nor can the individual curves be reconciled with any accuracy to an equation of the form of (1), which causes us to believe that the hyperbolic relation constitutes only an empirically approximate statement, but does not represent any physical law. Furthermore, none of the postulated steps would conform to such a relation.

Comparison with the earlier publication will show that the curves depart very widely in character from those corresponding to a migratory phenomenon. Attempts were made very unsuccessfully to reconcile them with the earlier equations by considering that in the prequenching treatment some graphite has been formed whose presence might be equivalent to the shifting of the time axis to a negative value. It is quite evident that this expedient will not in the slightest degree help to affect such a reconciliation. If the reaction were governed by the rate at which carbon could crystallize out of its solid solution, the rate of formation of free carbon would increase with time for the interface between carbon and solid solution obviously increases as the amount of carbon deposited increases. In the present instance this rate decreases with time, which rejects the theory of a reaction at the carbon interface, but is in favor of an explanation based on graphitization being controlled by an interfacial reaction at the boundary between cementite and solid solution. If such a relation exists, then under the circumstances now being investigated graphitizing rate is determined by the rate at which carbon (or cementite) can go into solution. It can be readily shown that if the rate of graphitization is so controlled then it is determined by the area of the interface at any given moment, and if the cementite grains are originally equiaxed the carbon available for graphitization, k, at time, t, is given by the equation

$$k = (a - bt)^3 \qquad (2)$$

if the cementite particles are rod-like in character, that is, have one dimension largely in excess of the others, the equation will be

$$k = (a - bt)^2 \qquad (3)$$

and if they are in thin plates having two dimensions approximately equal, the third being negligibly small compared with the other two, the equation will be

$$k = a - bt \qquad (4)$$

Now it is impossible to pass an equation of any one of these three forms satisfactorily close to the observed points in Figs. 3, 4, 5, and 6. We can, however, compute an equation of the form (2), which will agree very closely with the observed facts for the earlier portions of the four curves, and these equations are shown in connection with the corresponding loci in the four figures. The observed facts, however, depart from these loci long before k has become zero. This departure may be due to any of several causes.

If all the cementite grains are equiaxed, but the individual grains vary considerably in dimensions, then the time will come when the smallest grains will have disappeared, while the larger ones survive. The reduction in the value of k will then proceed at a slower rate than given by equation (2), to the extent that some of the interfaces are no longer existent. Under the effect of an interfacial reaction the radius of the particles will decrease at a uniform rate with time or conversely the time for the complete destruction of a given particle is inversely as its radius. Graphitization is complete when the largest particle has been completely destroyed (omitting a correction for the graphitization of carbon, which is soluble as austenite, but not as boydenite). The time at which graphitization will end, therefore, should on these assumptions bear to the time at which the smallest particle is destroyed the same ratio that the radius of the largest particle bears to the radius of the smallest particle.

An inspection of the curves shows that this ratio is extremely large, approaching on the average to infinity, and it is somewhat unlikely metallographically that there be so vast a difference in the size of the cementite particles. The form of the curve between the time when it departs from the theoretical cubic function and the time when k becomes zero would depend upon the distribution of sizes of cementite particles which distribution might be in accordance with the probability law. In the absence of knowledge as to this distribution we are unable to approximate a correction by which we might calculate the course of the curve between the two points. If, on the other hand, we make an assumption which is somewhat consistent with the metallographic appearance of the carbides in such samples that some of the cementite exists in the form of equiaxed grains and the rest in the form of relatively large rods, then in the early steps of the reaction the small equiaxed grains exposing a

relatively large surface will in the main control the rate of formation of graphite, but will be soon destroyed, leaving behind only the larger rods which have not yet been very strongly attacked. The form of the equation should then change during the progress of graphitization from that of (2) to that of (3).

On each of the four figures is plotted a quadratic equation which, although it is not a perfect square, at least approximates that condition and agrees very closely with the latter portion of the observed graphitization curve and which, furthermore, passes not so very far from the point at which the cubic curve diverges from the observed values. This is considered to be a fair degree of corroboration of the postulate that the conditions here observed correspond to the occurrence of an interfacial reaction, the geometry of the cementite being in accord with the assumptions just made. Such a geometry is, furthermore, in reasonable harmony with the dendritic freezing of white cast iron, the equiaxed grains of carbide represent land-locked masses of ledeburite, while the slender cylinders represent the filling between the branches of a given dendrite or of adjacent dendrites.

We may offer approximate confirmation of the assumption concerning the two forms of grains along the following lines. The equiaxed grains would be the first to dissolve on heating white cast iron which has been slowly cooled to a temperature above A_1 and also are the first to be destroyed in the graphitizing reaction, both conclusions arising out of the relatively small sizes of the individually small equiaxed grains and their large number and aggregate surface. At the point where the observed loci depart from the cubic equations there should, irrespective of temperature, have been destroyed the same amount of carbon, since in all cases the amount of carbon present in the form of small spheroids should have been the same in all the samples. Therefore, at this point of departure the decrease in k observed, plus the solubility of carbon at the temperature of observation, should be the same. Actually they are approximately as shown in Table IV.

Table IV	
Temperature	Percentage of Carbon
1650 degrees Fahr. (900 degrees Cent.)	2.00
1515 degrees Fahr. (825 degrees Cent.)	1.96
1435 degrees Fahr. (780 degrees Cent.)	1.81
1290 degrees Fahr. (700 degrees Cent.)	1.30

With the exception of the last line the numbers are of the same order of magnitude and having in mind that they depend upon a reading of the point of tangency of two very nearly parallel lines, as well as upon some arbitrary choice as to the number of observed points at the beginning of each curve to which equation (2) is to be fitted by the method of least squares, it is thought that this agreement is sufficiently close so as not to contradict the postulates.

We may consider also evidence with respect to the thermal coefficient of reaction velocity. It is quite obvious that cementite cannot dissolve into the surrounding solid solution any faster than it is carried away by migration, for otherwise there will be built up next to the cementite a layer of solid solution saturated with respect to the carbide into which no more carbon could pass. We observe the thermal coefficient of the present reaction's velocities to be about 1.04. Ishiwara[8] has observed the thermal coefficient of the rate of migration of carbon in iron to be about 1.103, the values are of the same order of magnitude, and both are less than the thermal coefficient of the graphitizing reaction when controlled by the migratory rate. We may point out in this connection that when graphitization is determined by rate of migration the migratory rate of carbon in iron is not the only factor which is a function of temperature. The concentration gradient decreases with temperature to the extent that the A_{cm} and the A_{gr} lines approach one another, according to our present conceptions, as the temperature rises. Furthermore, the number of graphite nodules is generally lower when graphitization progresses at low temperature than at high. This point has never been quantitatively demonstrable, but at least there is qualitative evidence in this direction. The fact that the thermal coefficient of graphitizing reaction when controlled by rate of migration is 1.23 and, hence, greater than the thermal coefficient of the diffusion rate indicates that the reaction is accelerated by temperature more rapidly than is the mere diffusion rate, for the reason that the migratory paths are decreased in length, due to an increased number of nodules, and this acceleration is more than sufficient to balance the decreased concentration gradient arising out of the changed differences in solubility.

Two further points should be noted before closing. The

[8]Ishiwara, "Effect of Impurities on the Dendritic Structure in Carbon Steels and Their Diffusion at High Temperatures," Science Reports of the Tohoku Imperial University, Sendai, Japan, Series I, Vol. XII, No. 4, 1924, page 327.

graphitizing rates under the present conditions are very much faster than under the migratory conditions. Yet, the migratory rate originally the slowest reaction in the sequence is now faster than the solution rate. This means not that the interfacial reaction has been slowed up, but that the migratory rate has been accelerated. In part this may be due to shorter migratory distances for the prequenched material generally shows more and smaller carbon nodules than the normal. The present authors doubt, however, whether this is the entire explanation, although they are not able to offer further details. Also the thermal coefficients of the two types of reaction are distinctly different and it is quite possible that conditions might exist under which at high temperature the interfacial reaction may control and at low temperatures the migratory reaction. The former is the slower under the conditions here described, but it decreases in rate less rapidly with temperature than does the migratory reaction. The latter might, therefore, be the faster at one temperature, but being retarded more than the former, as the temperature falls, may at lower temperatures become the determining factor. This is not experimentally true in the temperature range here investigated solely because of the quantitative relations between the various factors involved.

In closing we must for emphasis repeat that the relations herein described are not applicable to metal cooled from temperatures above A_1 at less than some minimum rate, the precise velocity not having been studied in this paper.

DISCUSSION

Written Discussion: By permission of Dr. John White, Professor of Chemical Engineering, Rose Polytechnic Institute, Terre Haute, Ind., the following verbatim extract from a thesis by Emil A. Krockenberger presented under the title of "Investigation of Electrolytic Isolated Cementite" presented in partial fulfillment of the requirements for the degree of Bachelor of Science in Chemical Engineering is presented here as a dis-

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Note:—Since the preceding was written nodule counts on the material have been made, showing that the number of nodules of temper carbon, per unit volume, in metal malleableized after prequenching, is nearly independent of time and about one thousand times as great as in normal iron.

Computation has shown that under these circumstances the diffusivity of carbon in iron might be as low as ¼ the value calculated from the graphitizing rate of normal iron without retarding the interface reaction. The increased number of graphite nuclei thus seems to be the determining factor in the accelerated graphitization.

cussion of the paper "Graphitization of Prequenched White Cast Iron"
by H. A. Schwartz, H. H. Johnson, and C. H. Junge.

The data and the curve relationship of the carbon content and current
density of electrolytically separated carbides obtained at Rose Polytechnic
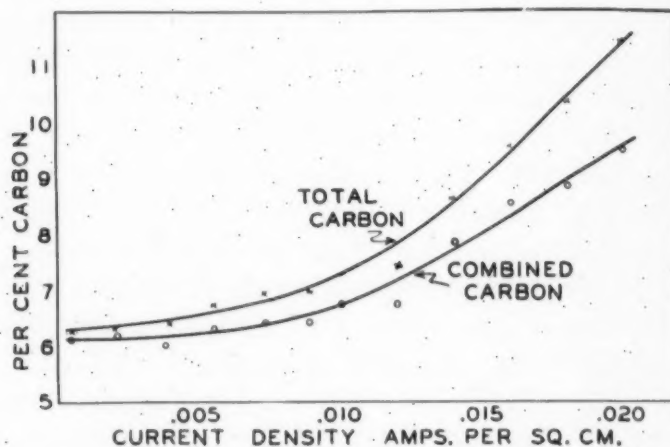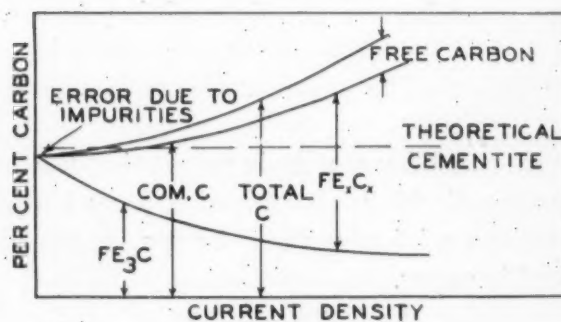

Fig. 1

Fig. 2

Institute are shown in Figs. 1 and 2 and Table I. This data shows that
the higher the current density the larger the percentage of both free
and combined carbon in the product.

The higher current densities formed residues which expanded and
formed longitudinal cracks with part of the product scaling off. Practi-
cally no outward change in appearance occurred in the lower current
products. The low current carbides were more metallic while the high
were more amorphous.

The carbides when brought in contact with water gave off a gas
resembling acetylene in odor, but tests with cuprous chloride in ammonia
gave negative tests for acetylene. The gas is probably some homologue
of acetylene. High current products gave off more gas than the low
products did.

After boiling with water high carbon residues gave lower carbon
analysis than before boiling.

Left margin notes:
Iron"
current
technic

Table I
Results Using C. P. HCL

| No. | I | Total Length | | Voltage | | Electrode Immersed (cm.) | Current Density (Amp./□ cm.) | $\frac{\Delta I}{T}$ (Milliamps) | $\frac{\Delta r}{T}$ (cm.) | Hr/.01 cm. | Carbon | | |
		Inches	cm.	High	Low						Total	Free	Combined
1	1.35	5.3	13.5	1.00	0.58	7.3	0.020	8.1	0.00476	21	11.41	1.90	9.51
2	1.22	5.5	14.0	0.86	0.52	7.0	0.018	6.8	0.00429	24	10.34	1.49	8.85
3	1.09	5.4	13.7	0.57	0.46	6.3	0.016	5.2	0.00381	27	9.53	1.00	8.53
4	0.96	5.2	13.2	0.73	0.44	6.8	0.014	3.9	0.00333	30	8.60	0.95	7.85
5	0.83	5.2	13.2	0.64	0.44	6.5	0.012	2.8	0.00286	35	7.41	0.54	6.73
6	0.70	5.5	14.0	0.53	0.44	6.0	0.010	2.1	0.00238	42	7.23	0.50	6.73
7	0.59	5.3	13.5	0.78	0.44	7.1	0.0088	1.57	0.00209	48	6.97	0.51	6.46
8	0.48	5.2	13.2	0.77	0.43	6.2	0.0073	1.06	0.00174	58	6.95	0.43	6.42
9	0.37	5.3	13.5	0.48	0.38	6.7	0.0055	0.61	0.00131	77	6.74	0.39	6.35
10	0.26	5.4	13.7	0.56	0.40	6.3	0.0038	0.29	0.00090	111	6.43	0.38	6.05
11	0.15	5.5	14.0	0.45	0.31	7.0	0.0021	0.09	0.00050	200	6.32	0.13	6.19
12	0.04	5.4	13.7	0.46	0.24	6.0	0.00058	0.007	0.00014	700	6.30	0.14	6.16

When specimens from high current densities were pulverized in contact with water considerable heat was given off, and the carbides corroded very readily.

Voltage did not affect the results very noticeably except when it changed the current density:

Applied Current	Voltage	Combined Carbon
0.05 amp.	0.34	5.89%
0.05 amp.	8.90	6.01%

Lower current carbides were harder and appeared more like cementite should appear than the higher did.

Qualitative graphs of the effect current density has on certain chemical and physical properties of the resulting iron carbide are shown in Fig. 3.

Microscopic examination showed that the lower current products were composed almost entirely of pure cementite for it showed the true structure of cementite surrounded with voids. Slightly higher current products showed the outlines of the cementite islands with dark spots on them, and very high current products only gave a few small cementite spots.

During the electrolysis the peculiar gas mentioned above was given off.

The lower current carbides had a higher density than the higher did.

From a study of the carbon content microscopic examinations and the physical and chemical properties of the residues at different current densities and voltages we believe the following:

The constituents in the residues are Fe_3C, some other iron carbide which is very active chemically, free carbon, oxygen, and impurities that were in the iron before electrolysis, such as silicon, sulphur, etc. The
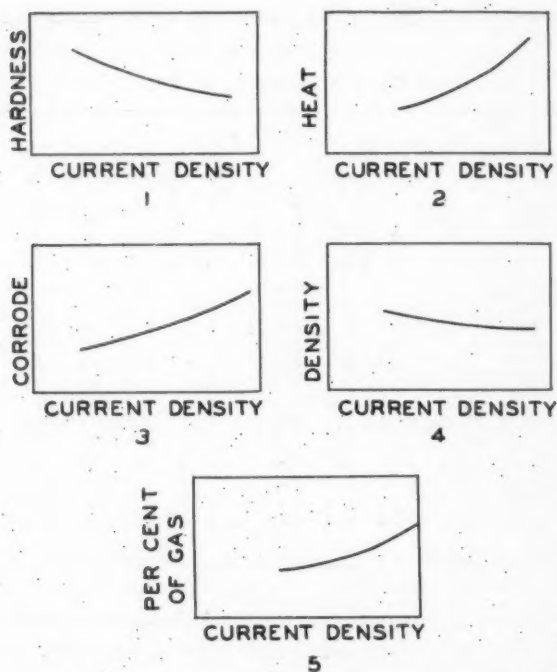
Fig. 3—Curves Showing Chemical and Physical Properties of Resulting Iron Carbide. (1) Hardness of Carbide. (2) Ability to Liberate Heat When Pulverized with Moisture. (3) Tendency to Corrode. (4) Density of Carbide. (5) Ability to Liberate a Gas When Put in Water.

amount of each depending on the current density used. The oxygen is due to double decomposition between the active carbide and water,

$$Fe_x + H_2O = Fe_xO_x + C_xH_x$$

The active carbide is formed by a decomposition of cementite and was not in the iron before electrolysis.

The gas given off is due to the above equation.

The decomposition phenomena is indirectly a function of the current density rather than the external voltage. From Ohm's law, $E = IR$, the current density is the only function that will vary the electrical potential between the surface of the iron and the film of the electrolyte, or will vary the force tending to decompose the cementite. If the electrolysis could be carried on with a negligible electric current so that no potential would exist, theoretical cementite would be obtained.

The approximate relation existing between the combined carbon and the current density (in the range studied) is:

Per cent Carbon $= 6.41 + 11.4$ CD $- 2067$ CD^2
(By least squares)

We believe that the action taking place is a decomposition of cementite into a higher carbide and finally into free carbon, and that the

degree of decomposition varies with the electrical potential existing between the surface of the iron and the film of the electrolyte, the potential being a function of the current density and not the external voltage.

Written Discussion: By R. M. Allen, American Car and Foundry Company, New York City.

The authors' method of attack of this interesting problem is quite original and the results of their investigation appear conclusive.

A high power microscopical study of the nature of chilled iron and the mechanism of graphite precipitation carried on sometime ago, appears to throw additional light on the subject. This method of investigation offers another method of attack and will yield promising results when combined with methods such as those employed by the authors of this paper.

In considering the question of the graphitization of prequenched white cast iron, in the absence of any proof to the contrary, it should be assumed that the precipitation of graphite follows the same fundamental laws that operate in the case of the precipitation of graphite in ordinary gray iron.

The theory of the presence of graphite nuclei upon which the larger graphic masses grow during the period of temperature range which is conducive to graphite precipitation, is almost universally accepted as satisfactorily explaining the phenomenon. One significant fact, however, cannot be so easily explained on the basis of this theory, although it does not necessarily deny it. If these nuclei actually exist in the molten magma as carbon atoms properly grouped together to serve as nuclei, it is logical to assume that they are more or less uniformly dispersed throughout the entire mass and therefore in cooling, the graphitization should be capable of starting at any point where they exist. An examination of chilled gray iron reveals, however, the fact that graphite precipitation starts only and always in the pearlitic areas of the dendrites and never in the cementite matrix. Micrograph, Fig. 4, which is a general view of chilled iron at a magnification of 200 diameters will serve to illustrate this point. It will be seen that graphite precipitation has started in this chill but in every case the graphite is located within a pearlitic area and that the white areas (the so-called "cementite") are entirely free of graphite.

It appears that something, other than the presence of graphite nuclei is necessary before the cementite can effect the precipitation of graphite. Fig. 5 shows the appearance of chilled iron at a magnification of 3000 diameters. Here we see a very early stage in the formation of the dendritic portion of the chill which indicates that the mechanism of formation of the dendrite itself is complex. The first material that forms in the dendrite differs but very little in microscopical appearance from the cementite matrix, but is always revealed as distinct from it by etching with a boundary line between the two. This material (austenite?) then proceeds to decompose into graphite and pearlite. If the rapidity of chilling is sufficiently great, the formation of well defined graphite is prevented, but some pearlite will be formed. In the present case, however, even the latter is but poorly developed. Some darker areas can be
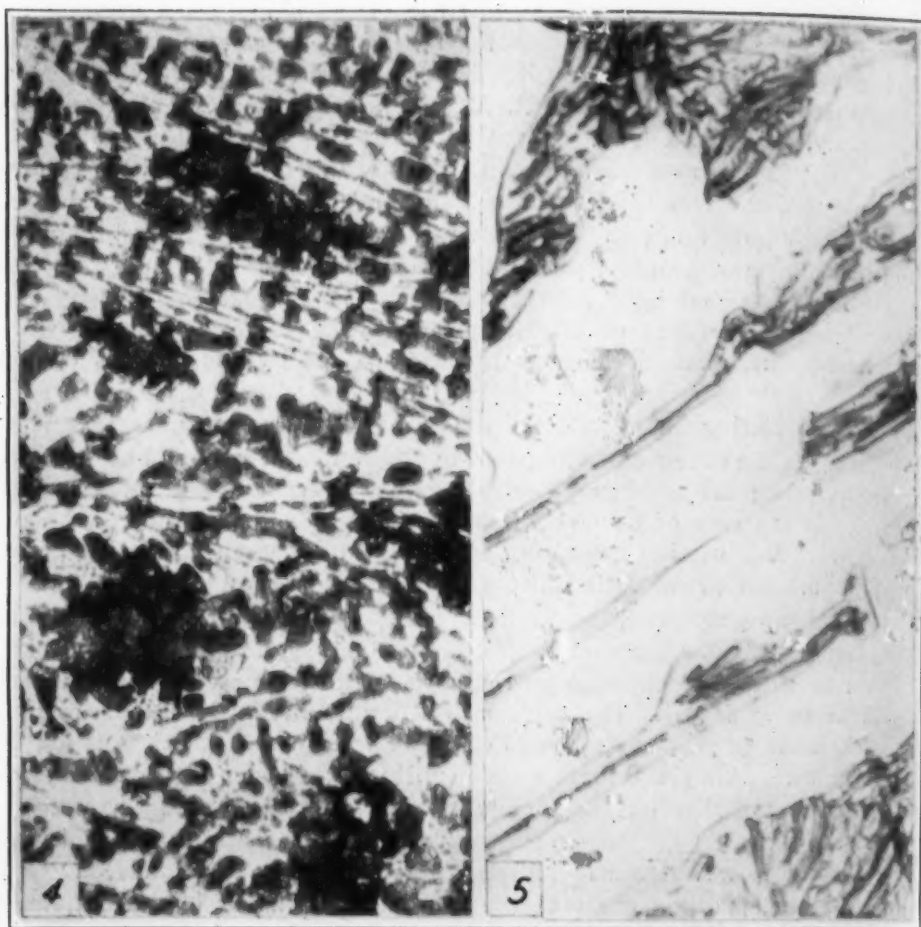
Fig. 4—Photomicrograph of Graphite in Chill, Showing Occurrence Only in Pearlite Areas. × 200.
Fig. 5—Photomicrograph of Chilled Cast Iron, Etched with Nital. × 3000.

seen within the pearlite. These indicate an incipient stage in the formation of the graphite and they can be traced through various stages of development up to the complete graphite flake.

Fig. 6 (2000 diameters) shows a very early, but more advanced stage than Fig. 5, of the formation of graphite within the pearlite. If the temperature range is such that the graphite can continue forming, the tendency is for these small graphite precipitations to merge and form large masses. Such additional graphite as may be formed, will, of course, be at the expense of the cementite plates and the end products would be graphite and ferrite.

Fig. 7 (3000 diameters), which is from a cast iron having sufficient nickel and chromium to stabilize the martensitic condition at the temperature where pearlite should ordinarily form, shows a still more complete formation of the graphite flakes, but it will be evident that even here
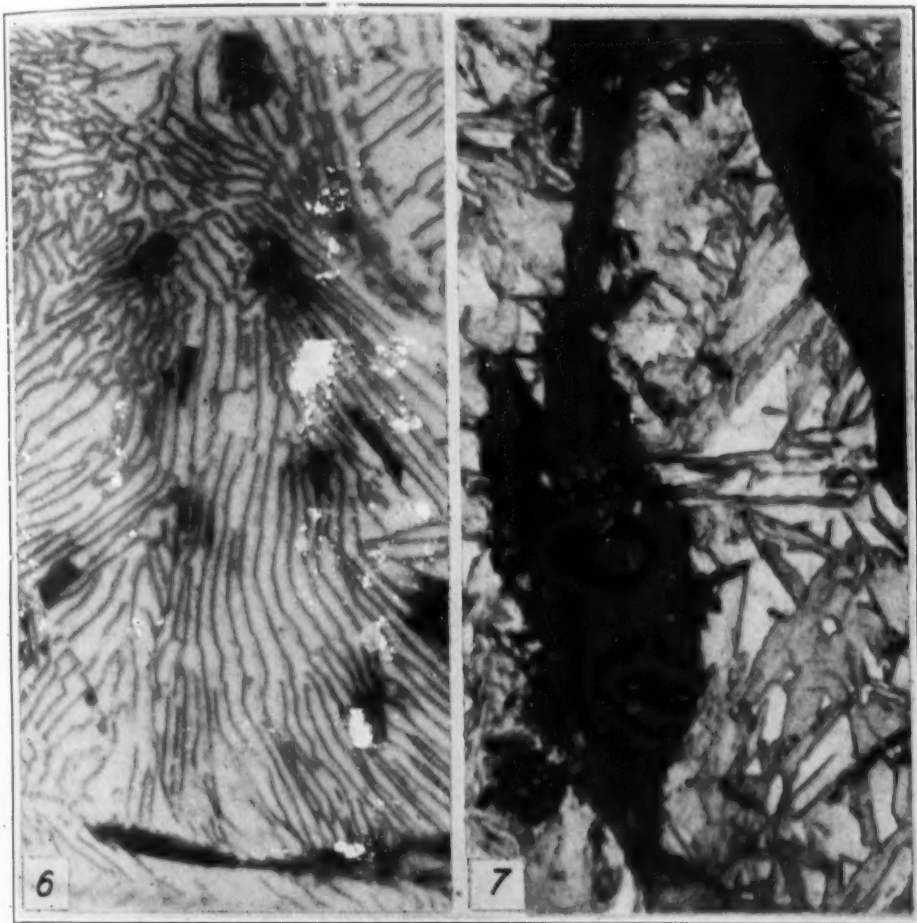
Fig. 6—Photomicrograph of Early Stage of Graphite Precipitation in Cast Iron. $\times$ 2000.
Fig. 7—Photomicrograph of Graphite in Nickel-Chromium Cast Iron of Martensitic Structure. $\times$ 3000.

development is not complete, and that streamers of graphite are concentrating from the outside areas into the larger masses.

When we examine chilled cast iron at high power after a suitable etch such as alkaline sodium picrate we obtain additional light on what is taking place. The sodium picrate etch has the property of staining pure iron carbide a deep black. In the case of the so-called cementite matrix, we do not obtain a black stain but only varying shades of brown. Imbedded within this, however, are numerous small bodies which stain very deeply.

Fig. 8 (3000 diameters) shows the presence of a considerable number of these small bodies. In considering them it will help to realize somewhat their size and nature when it is pointed out that they are only a fraction of the size of most of our common pathological bacteria such as diphtheria, typhoid, etc. It is quite evident from their reaction to sodium picrate,
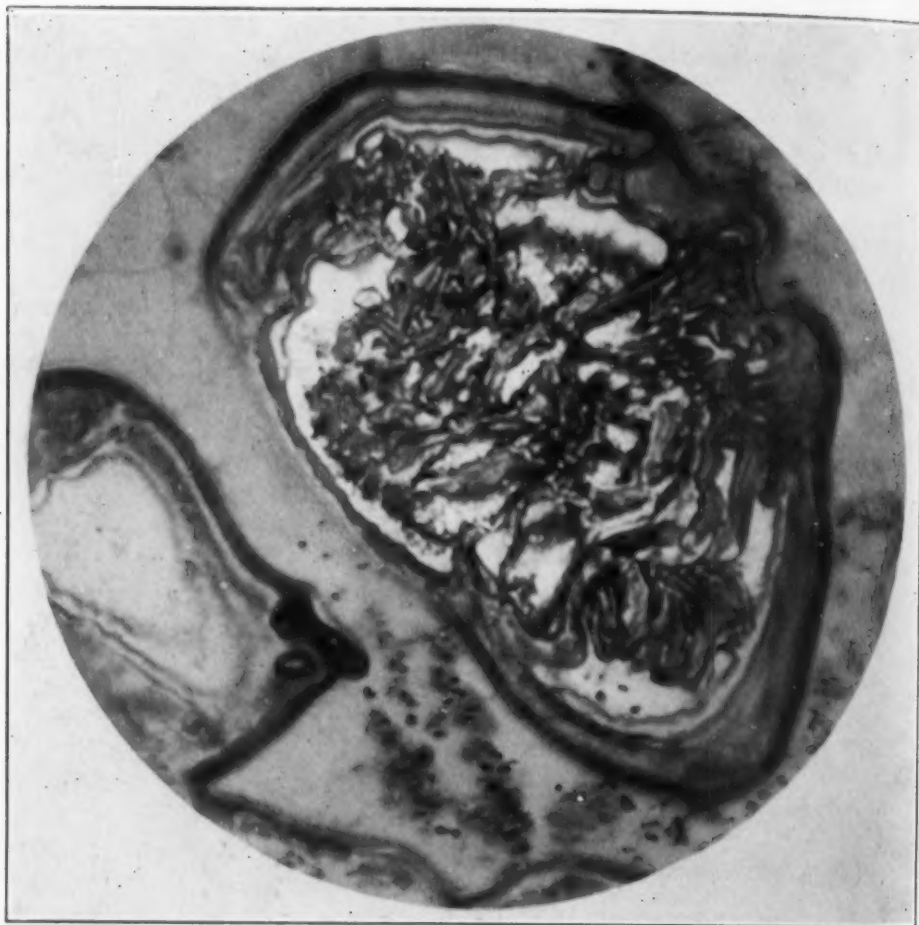
Fig. 8—Photomicrograph of Chilled Cast Iron, Etched with Sodium Picrate, Showing Concentration of Fe_3C in "Cementite" Areas.　$\times$ 2600.

that these are not carbon nuclei but iron carbide Fe_3C (or some other carbide).

Another condition which should be noted is the wavelike concentration of carbide occurring in the dendritic area which can be seen as multiple bands, the outermost band being the deepest in color. The innermost band merges into the white field which is nearly pure ferrite. Thus it can be demonstrated that iron carbide exists in varying concentrations in chilled iron. In other words, chilled iron is not in a condition of equilibrium.

The authors of this paper have shown that the graphitizing rate is determined by the rate at which carbon or cementite can go into solution and from their equations have proved that the nature of the cementite particles plays an important part in graphite precipitation.

While I have made no investigation of prequenched samples, a study of the low power micrographs, Figs. 1 and 2 of their paper indicates

that what has taken place has been the concentration of a much larger number of carbide or cementite particles and that they are probably of a considerably larger size than shown in my Fig. 8. Whether or not graphite nuclei are present it would appear that the primary cause of the increased rate of graphitization with prequenched iron is due primarily not to original graphite nuclei, but to an increase in the number of carbide particles (or concentrations) which then dissociate to form graphite nuclei and ferrite in substantially the same manner that graphite is formed during the formation of the pearlite in the dendrites.

DR. ANSON HAYES:[1] Sometime ago I received a copy of this very interesting paper with the request that I prepare a written discussion and submit it by September 1. There is a difference between delving into the mysteries of making ingot iron and those of making white iron for castings. It is so great and so time-consuming that I have not found time to prepare anything in the way of a systematic discussion of the paper, but I have jotted down, since I have been here, a few scattered thoughts in regard to the subject. The influence of quenching on the rates of graphitization of white cast iron below the critical range is one which we studied at quite some length during the years 1922 and 1923. We are fully in agreement with the general conclusions that the increased rate of graphitization is due to greater numbers of nuclei with consequent greater surface exposure for the surface reaction which is involved in the breakdown of the carbide. And further, that the shorter distances over which migration must take place also is a contributing factor to the increased rate of graphitization. In regard to the observations which have been recited as to the composition of the products obtained by electrolyzing iron-carbon alloys and in regard to the questions raised as to the existence of a pure carbide, perhaps it will be of assistance to the author and to those who are interested in carrying investigations further in this regard to recite some of our experiences in the laboratory at Iowa State College. Dr. George Brody, and Dr. William H. Jennings in the work of the heat of formation of cementite which was published in the TRANSACTIONS of the A.S.S.T.[2] attempted to prepare their cementite from Armco iron and pure graphite by melting the iron in contact with an excess of graphite. The details of this method need not be repeated here, but it is of interest to know that they obtained a product which showed 16 per cent carbon. Magnetic separations following successive grindings to a fineness reaching 300 mesh failed to bring about any further separations of graphite and carbide. At one time these men had me almost convinced that possibly they had discovered a new carbide. In the course of running this thing down we went to a method of purification of the carbide which is suggested by one of the earlier German workers in which the liquid acetylene tetrobromide (which has a density greater than that of graphite and less than that of carbide) had been used and these gentlemen went to work with this liquid in an attempt to make a separation.

[1] Member A.S.S.T. Director of Research, American Rolling Mills Co., Middletown, Ohio.

[2] G. H. Brodie, W. H. Jennings and Anson Hayes, "Heat of Formation of Cementite as Electrolyzed from Pure Iron Carbon Alloy of Eutectoid Structure and Composition," TRANSACTIONS, American Society for Steel Treating, Vol. 10, October 1926, p. 615.

In a short time they had one fraction which analyzed 85 per cent graphite and another fraction which analyzed a little more than 8 per cent graphite, and as we were interested in preparing pure carbide rather than any other features of investigations which might have been carried out on this material we immediately drew the conclusion that we were dealing with a mixture of graphite and carbide and immediately chose another method of getting our carbide. We then went to the preparation of eutectoid alloy of pure iron and pure carbon and had no difficulty in obtaining by electrolytic methods which are fully described in previous literature a carbide which showed almost exactly theoretical carbon content. Previous to this work we have used a pure iron-carbon alloy and an equilibrium method of study to determine the free energies of formation of iron carbide at the temperature 650 and 700 degrees Cent. From these two values of the free energy formation of carbide we calculated a heat of formation of minus 19,200. Two investigators previously made a determination of the heat of formation of iron carbide by a direct calorimetric method in which they had used a pure iron-carbon alloy prepared by melting the alloy in the presence of excess carbon. They did not describe in detail the methods which they used to purify their carbide and it was due to the statements which had been made in a paper they wrote that Dr. Brody and Dr. Jennings went to the methods they did in preparing their initial alloy in which they thought for a time they had obtained a carbide containing 16 per cent carbon. In regard to the possibility of a difference in composition of eutectoid carbides I wish to present further information. In a work which Dr. Brody and Dr. Jennings did they obtained the value of minus 13,850 calories for the heat of formation of carbide by the direct calorimetric method. This we were comparing to the value of Ruff and his workers of minus 15,300 calories. The difficulty with the calorimetric method, of course, is that we must assume some initial and some final state for the carbide and the oxides that are produced. When we had corrected Ruff's values by assuming the same heats of formation of the oxides involved as we had used in the latter determination, Ruff's value and the one obtained by Dr. Brody and Dr. Jennings were within some 400 calories of each other. I wish to point out the fact that Ruff's carbide was produced from an impure iron-carbon alloy of the following composition

	Per Cent
Carbon	4.13
Manganese	0.15
Phosphorus	0.20
Sulphur	0.006
Silicon	0.074
Copper	0.005

by melting this alloy in contact with an excess of lampblack. He must have gotten eutectic carbide under these conditions, and yet the analysis of his material conformed almost exactly to the theoretical compositions for the carbide, and the heat of formation value which he got fell well within the two values which workers in my own laboratory obtained.

Reply to Discussion

The authors are grateful to those who have given sufficient time and thought to our problem to prepare discussions.

We have been greatly interested in the suggestion contained in Mr. Krockenberger's thesis as to the reactions leading to the formation of a high carbon carbide in the electrolysis. The subject is one meriting thorough investigation. Since reading Mr. Krockenberger's results, analyses have been made of the gases resulting from the electrolysis. It may be said in brief that these gases are substantially hydrogen containing a very little hydrocarbon of some kind; the proportion proved insufficient to permit of determination by the ordinary methods of gas analysis but by combustion methods the electrolytic gases were shown to contain somewhat under a milligram of carbon per liter of hydrogen. Some slight evidence was also obtained that a liquid hydrocarbon accumulated in solution or perhaps colloidal suspension in the electrolyte. In neither case, however, was it possible to identify the compounds present.

In commenting on Dr. Hayes' discussion it will be necessary to invite attention to the fact that our paper concerned itself with the agraphitic, i. e., nongraphitic carbon content of the cementite. We are aware of the possibility of the formation of free carbon but determined, by the usual methods of analysis, the apparent graphitic content of the cementites and deducted this amount from the observed total carbon.

Personal correspondence with Dr. Hayes develops the fact that he suspects the presence of "graphitic acid", a compound which might be soluble in the course of an ordinary graphite determination and hence count as "combined" carbon though actually not involving a combination of iron and carbon. It was apparently such a compound as this that Jennings and Brody sought to separate by floating on a heavy liquid.

The assumption is tenable, in the absence of a direct experimental knowledge, and we owe thanks to Dr. Hayes for bringing this possibility to attention. Experiments are in progress looking toward a definite answer as to the possible existence of such material in our products.

The body of our paper gives reasons why we cannot agree with Dr. Hayes that the increase of surface exposed by graphite is the main cause of the acceleration of graphitization. In brief we do not agree with this for the reason that the observed graphitization temperature curves are not of a form consistent with Dr. Hayes' suggestion.

We are particularly interested in Mr. Allen's contribution since his photomicrographs show what may be considered to be graphitic nuclei in a very early stage of development. In the same connection we might allude to the fact that a graphite nucleus need not be graphite but may be some other material upon which the graphite can readily crystallize. We have no experimental knowledge of any such instances but consider that the possibility thereof exists.

STEEL AT ELEVATED TEMPERATURES

By Albert Sauveur

Abstract

Some of the physical properties at different temperatures of carbon steels, of austenitic and of nonaustenitic steels have been investigated by twisting small grooved bars. The factor of stiffness obtained by dividing the breaking load in pounds by the angle of twist in degrees bring out sharply the blue heat range in some steels. Time-strain curves are obtained which throw considerable light on the behavior of steel of different compositions when subjected to plastic deformation. Attention is called to an apparent relation between creep stresses and the torsional strengths.

THE first paper I ever wrote was presented to the American Institute of Mining Engineers in 1893 under the title "The Microstructure of Steel," a title familiar enough at present but at the time having very little significance for the great majority of metallurgists. I received on that occasion a letter from Edward de Mille Campbell, which I shall ask permission to read:

"Dear Sir:

"Having been interested for some time in the microstructure of iron and steel and especially in the paper that I heard you read at Chicago last summer, I have begun to have some of the work done here at the university under my direction. One difficulty that is always to be met in undertaking such work in a university is that of procuring suitable samples to represent the objects desired. I have obtained a good equipment for preparing specimens, polishing, etching, photographing, etc., and have already had a pretty good number of good mounts made, but I now want to get a few sections of rails, billets, etc., made or rather rolled at the proper heat and at too high heat, etc., to show if possible the effect of heat treatment described in your paper. Would it be asking too much of you to ask you to send me a few sections such as would in your opinion be suited to my purpose? If I can do so, I want to make the study of the microstructure of steel, etc., a regular part of the work that I have here in metallurgical chemistry. I feel the growing importance of the subject

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and hope to have it incorporated soon in our regular work in the chemistry of iron and steel.

"Hoping that you will not feel that I am asking too much of you, I am,

Ann Arbor, Michigan,   Very truly yours,
    February 14, 1894.   (Signed)   E. D. Campbell."

It is pleasant to recall as I am about to deliver, with full appreciation of the honor implied, the fourth Campbell memorial lecture, that I have been able in a small way to assist Professor Campbell in introducing at his university the study of metallography and in starting investigations which were to bring so much fruit.

Steel parts are frequently subjected in service to temperatures greatly exceeding room temperature and a knowledge of their behavior under these conditions is of much importance. The subject has attracted the early attention of metallurgists and engineers and the results of numerous and extensive investigations have been published in the proceedings of various scientific and technical societies as well as in book form. In the majority of cases these investigators have subjected test bars to the tensile test while maintaining them at some predetermined temperature, the yield point, tensile strength, elongation and reduction of area being duly recorded. Important results have been obtained. It has been conclusively shown for instance that carbon steel at least increases in strength and decreases in ductility in the so-called "blue heat range," generally situated between 250 and 400 degrees Cent. (480 and 750 degrees Fahr.). At higher temperatures the strength decreases rapidly while the ductility increases.

Tests of this kind were conducted in the metallurgical laboratory of Harvard University in 1922 and 1923, mostly by Dr. D. C. Lee, at the time a candidate for the doctor of science degree. They are mentioned here because they revealed more clearly, I believe, than had been done before, the discontinuities which occur in the tenacity, hardness and ductility curves, not only in the blue heat range but also at temperatures corresponding to the thermal points, A_1, A_2, A_3 and A_{3-2}.

The tensile strength and Brinell hardness of carbon steels at various temperatures between room temperature and 1000 degrees Cent. (1830 degrees Fahr.) are shown in Figs. 1 to 5.

A consideration of these results appeared at the time to justify the following conclusions.

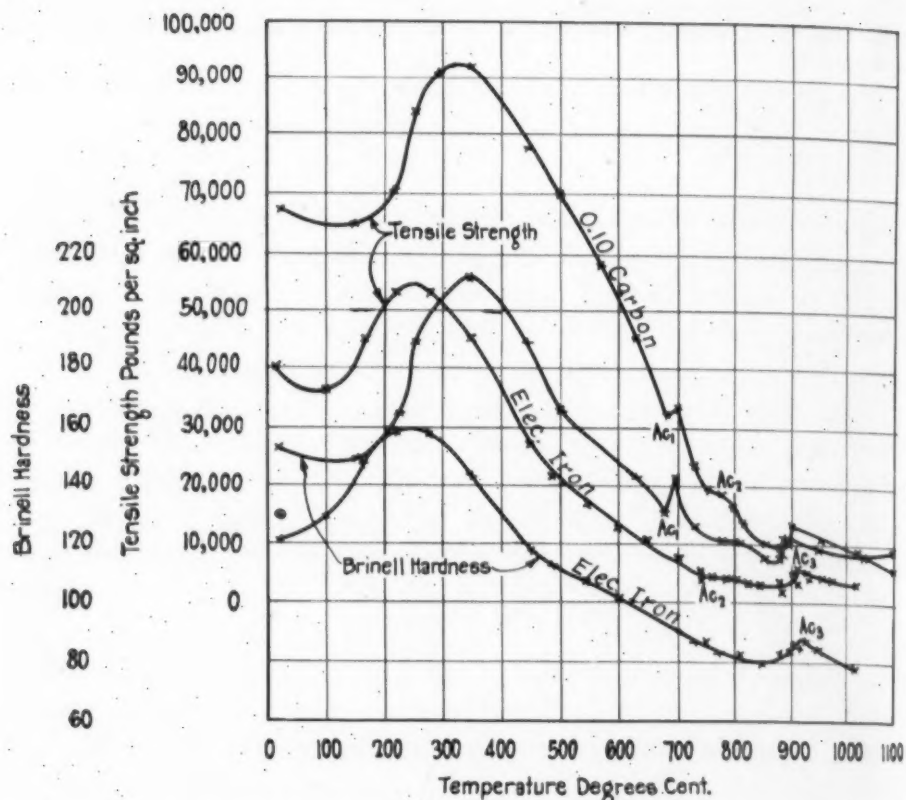1. On heating normalized iron and steel above atmospheric

Fig. 1—Tensile Strength and Brinell Hardness of Electrolytic Iron and 0.10 Per Cent Carbon Steel Between Room Temperature and 1000 Degrees Cent. (1830 Degrees Fahr.).

temperature their strength and hardness are at first slightly decreased, this being followed by an increase which reaches its maximum in the blue heat range of temperature.

2.   In pure iron the blue heat phenomenon occurs in the vicinity of 250 degrees Cent. (480 degrees Fahr.), with 0.10, 0.30, and 0.50 per cent carbon at about 325 degrees Cent. (620 degrees Fahr.), and with 0.75 per cent carbon near 400 degrees Cent. (750 degrees Fahr.).

3.   The increase of strength and hardness observed in the blue heat range probably results from the strengthening and hardening of the ferrite, little if any increase being detected in the absence of that constituent.

4.   Straining normalized iron or steel above its elastic limit increases both strength and hardness, and the increase is more marked as the temperature at which straining occurs rises above atmospheric temperature, being maximum after straining in the blue heat range.
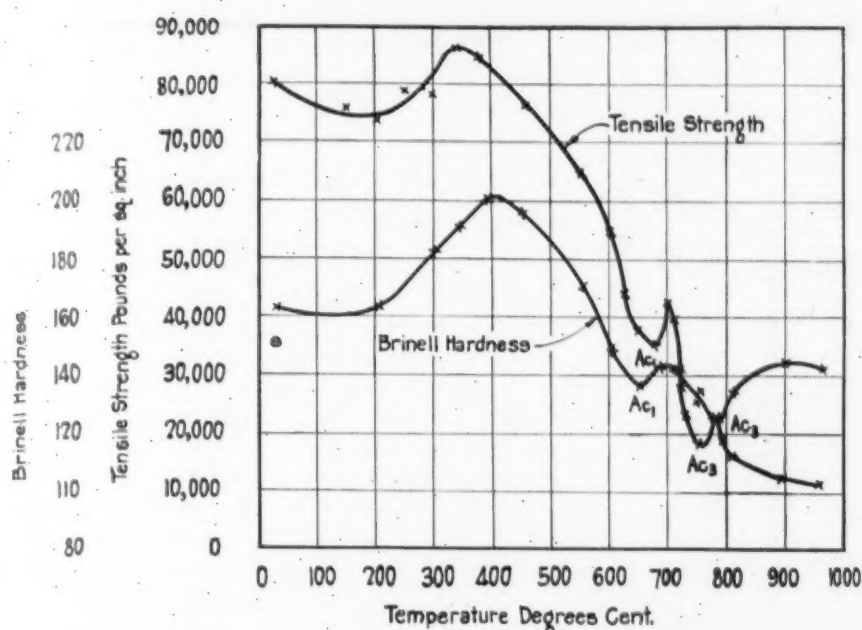
Fig. 2—Tensile Strength and Brinell Hardness of Steel Containing 0.3 Per Cent Carbon Between Room Temperature and 1000 Degrees Cent. (1830 Degrees Fahr.).

When the temperature exceeds some 580 to 680 degrees Cent. (1075 to 1255 degrees Fahr.) the straining of the metal results in decreased hardness. This range of temperature may be taken as the demarcation between cold working and hot working.

5. The hardness resulting from cold deformation at room temperature is increased further by heating the metal above atmospheric temperature, reaching a maximum after heating in the blue heat range, irrespective of the degree of deformation. To soften the strained metal it is necessary to heat it to a temperature exceeding 425 degrees Cent. (800 degrees Fahr.) for electrolytic iron, 500 degrees Cent. (930 degrees Fahr.) for steel containing 0.10 per cent carbon, and 600 degrees Cent. (1110 degrees Fahr.) for steel containing 0.50 per cent carbon.

These tensile tests at the higher temperatures clearly reveal discontinuities in the curves occurring at the thermal critical points. Similar results had been obtained by Rosenhain and Humfrey. (See Fig. 6.) This relation between the mechanical properties of steel and its thermal critical points does not seem to have received the attention it deserves.

The study of steel at elevated temperatures was again taken up in our laboratory some two years ago and the results so far obtained
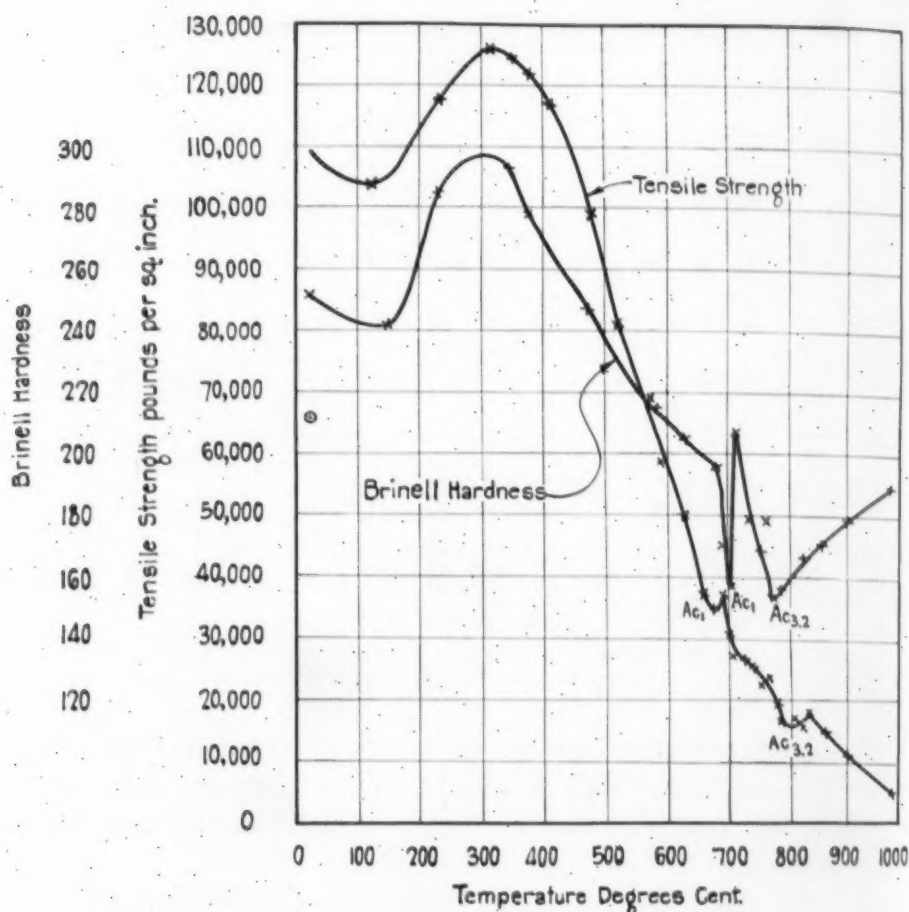
Fig. 3—Tensile Strength and Brinell Hardness of Steel Containing 0.50 Per Cent Carbon Between Room Temperature and 1000 Degrees Cent. (1830 Degrees Fahr.).

justify their presentation at this time. This paper can only be a report of the progress made and I refain from drawing any conclusion savoring of finality or dogmatism.

In these new experiments the twisting test was adopted, following in this in a general way, the method previously used by Dr. Lee and myself.[1]

The actual tests about to be described were performed for the most part by Ralph W. E. Leiter, a candidate for the doctor of science degree at Harvard University, and some by Dr. R. H. Aborn, instructor in metallurgy and metallography at the same institution. Much credit is due to their zeal, industry and skill, and it is a privilege to be permitted to discuss and interpret the results obtained by them.

[1] Albert Sauveur, "What is Steel?", first annual Henry Marion Howe Memorial Lecture, delivered at the New York Meeting of the American Institute of Mining and Metallurgical Engineers, February 19, 1924.
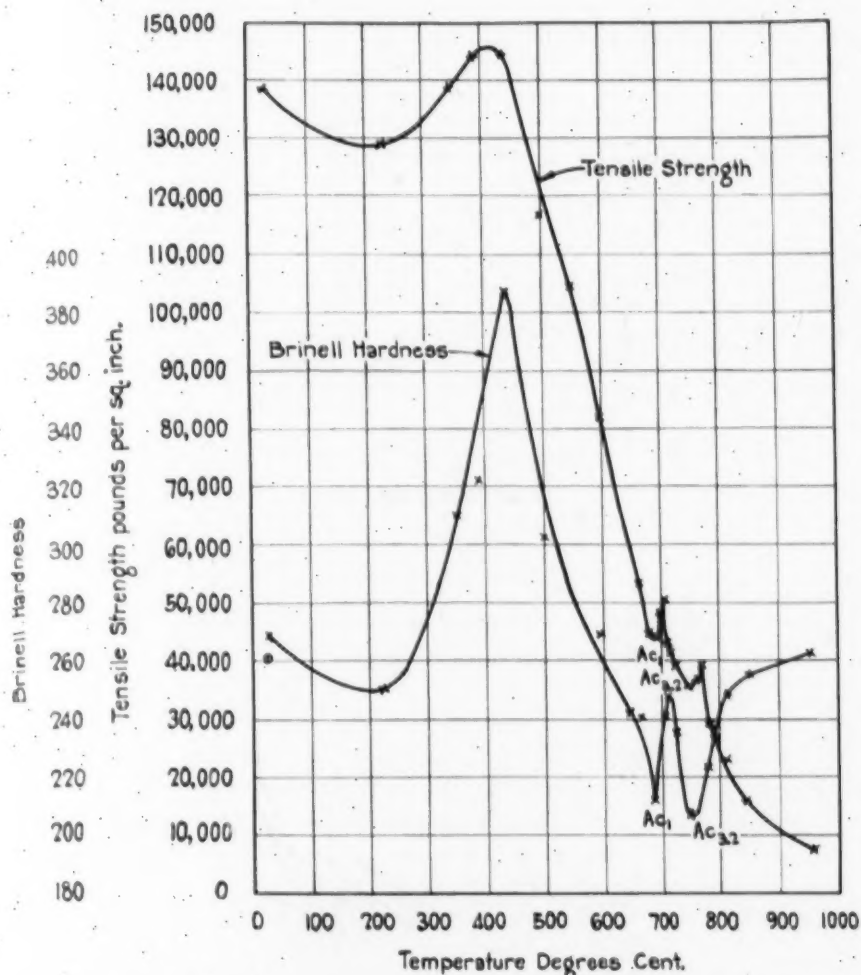
Fig. 4—Tensile Strength and Brinell Hardness of Steel Containing 0.75 Per Cent Carbon Between Room Temperature and 1000 Degrees Cent. (1830 Degrees Fahr.).

TEST BAR

The dimensions of the test bars used in these experiments are clearly indicated in Fig. 7. If preferable, round bars some 5/8 inch in diameter may be used. They should be squared at both ends, however, to facilitate gripping. Fillets have not been found necessary as the bars rarely break at the shoulders of the reduced sections.

TWISTING DEVICE

The twisting device is illustrated in Figs. 8 and 9. It will be seen to consist of an electrically heated resistance furnace, of suitable attachment for holding the test bar in such position that its reduced
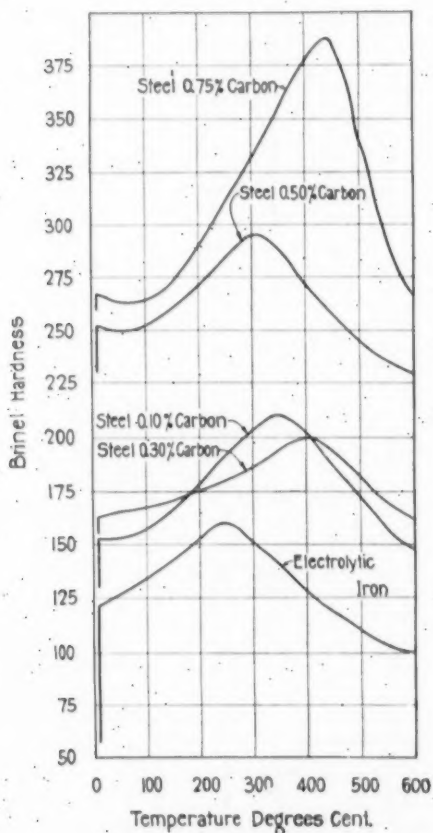
Fig. 5—Brinell Hardness of Electrolytic Iron and of Some Carbon Steels at Room Temperature after Straining at the Temperature Indicated.
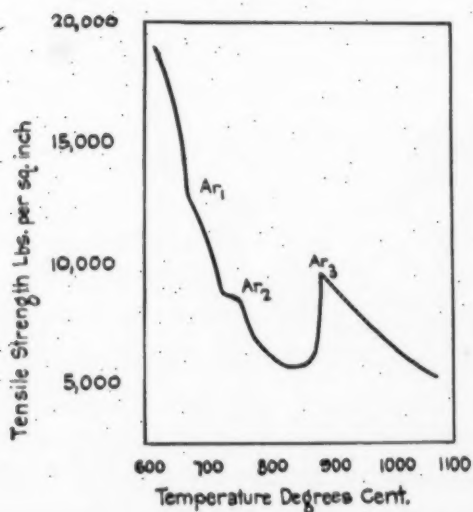

Fig. 6—Tensile Strength of Steel Containing 0.10 Per Cent Carbon Between Room Temperature and 1100 Degrees Cent. (2010 Degrees Fahr.) After Rosenhain and Humfrey.
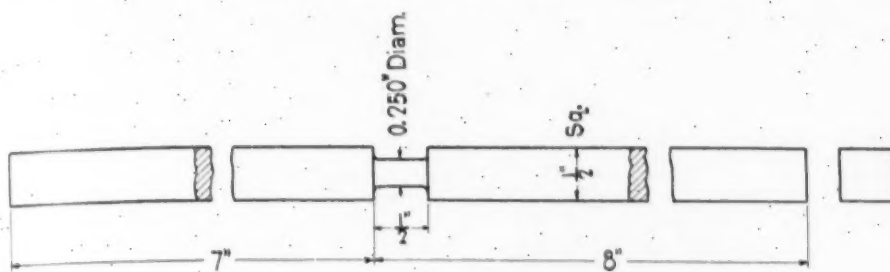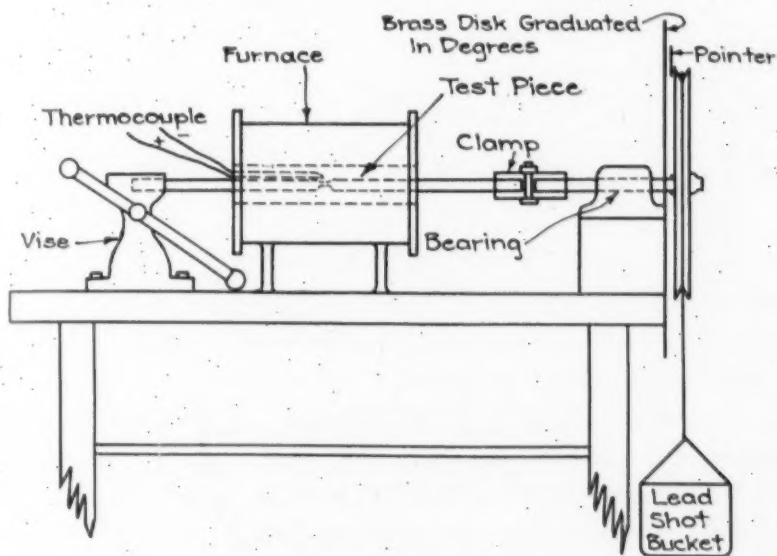
Fig. 7—Test Bar Used in this Investigation.

Fig. 8—Twisting Device Used in this Investigation. Vertical Section.

section occupies the middle of the furnace, of a pulley rigidly connected with the bar and to which is suspended by means of a cable, a bucket to receive the lead shot. A drum revolving by clockwork is so connected with the twisting pulley that diagrams are obtained in which the vertical distances are proportional to time and the horizontal distances to the twist. We have called these "time-strain" diagrams. The speed of the drum is thirty seconds per linear inch, while a complete trip of the pen, from right to left or from left to right, over the ruled paper wrapped around the drum corresponds to a twist of 180 degrees, a return trip indicating a complete revolution (360 degrees). As the travel of the pen covers about four inches each horizontal distance of one inch would correspond to a twist of 180/4 or 45 degrees, were it not for the fact that the pen travels faster as it moves towards the center of the drum and more slowly as it again moves away from it.
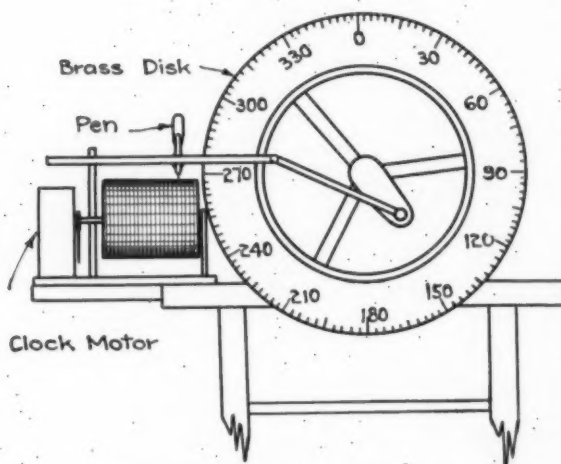
Fig. 9—Twisting Device Used in this Investigation. End View.

THE TESTING OPERATION

The test bar is inserted in the furnace and clamped rigidly at the twisting end only, so as to permit its free expansion during heating. The thermocouple is placed in close contact with the reduced section of the bar, the ends of the furnace plugged with asbestos and the heating of the furnace started. The bar is kept for thirty minutes at the desired temperature. The loose end of the bar is now solidly tightened in the vise and the loading bucket suspended from the pulley. The reading of the pointer is taken as zero for the particular test under way and the loading is begun by adding lead shot to the bucket, starting at the same time the rotation of the drum. After the bar has started to twist, it is endeavored to add the shot at such rate as to cause the bar to twist about 10 degrees in ten seconds, giving a full revolution in 360 seconds. When that rate of twist is reached it is maintained, if possible, until fracture takes place. It frequently happens, however, especially at elevated temperatures, that the twisting speed increases without further increase of load. This increase over the speed of the first revolution may be designated as the twist acceleration. If the twisting speed has been maintained constant at six minutes per revolution during the earlier part of the test, the difference between six minutes and the speed of the last twist may be taken as the twisting acceleration. It would be four minutes, for instance, if the last twist required but two minutes. At room temperature or at relatively low temperature increased twisting speed is generally quickly followed by fracture. It is believed

that this twisting acceleration varies in different steels and that it is one of their significant characteristics.

An arbitrary yield point value may be obtained by weighing the load corresponding to a small movement of the pointer such as 2 degrees beyond the zero reading.

The data obtained includes (1) the torsional strength expressed in pounds, that is, the actual weight required to break the bar, (2) the angle of rupture, that is, the number of twists or fraction of a twist, (3) the "factor of stiffness" obtained by dividing the stress by the strain, i. e., the breaking load by the twist, and (4) time-strain diagrams.

All test bars were exactly the same size and the conditions of testing did not vary, therefore the results should be truly comparative.

The blue heat brittleness is generally assumed to occur at the temperature for which the strength is maximum or the ductility minimum, although the two points do not always correspond. Considering that blue heat brittleness is probably directly proportional to the stress and inversely proportional to the strain it seems more logical to consider the ratio of the stress to the strain as a measure of its intensity. We have called this ratio the "factor of stiffness," and it will be shown that the occurrence of the blue heat phenomenon becomes by this method more pronounced, and that it corresponds to sharper peaks in the curves. Other discontinuities in the curves, to which attention will be called, undoubtedly have their significance also.

Ingot iron. The behavior of Armco iron at elevated temperatures under the testing conditions outlined above is shown in Fig. 10.

The three curves here shown put the blue heat phenomenon in the vicinity of 300 degrees Cent. (570 degrees Fahr.), but the peak in the factor of stiffness curve is much more pronounced and leaves little doubt as to the approximate position of that point, the exact location of which might, of course, be slightly to the right or to the left of the point indicating a temperature of 300 degrees Cent. (570 degrees Fahr.), that is, testing within closer ranges of temperature might cause a slight shifting in its position. The marked decrease of strength as well as of stiffness once the temperature of 300 degrees Cent. (570 degrees Fahr.) is passed should be noted. The three curves exhibit also sharp discontinuities in the vicinity of 800 degrees Cent. (1470 degrees Fahr.) and 875 degrees Cent. (1605 degrees Fahr.), that is, at or near the A_2 and A_3 points. It is believed that
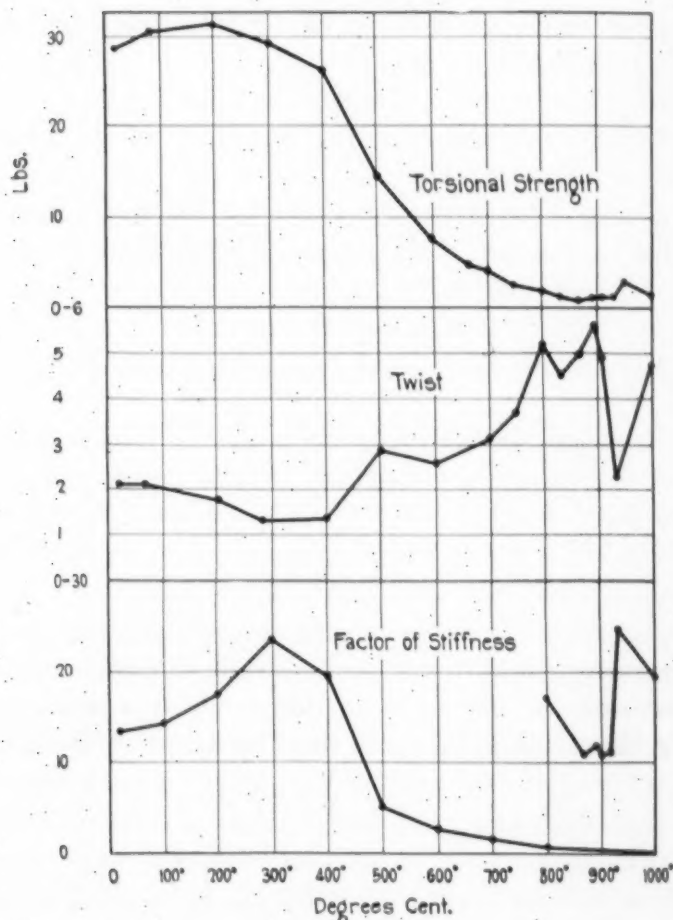
Fig. 10—Twisting Tests of Armco Ingot Iron.

from room temperature to 900 degrees Cent. (1650 degrees Fahr.) the curves reveal the behavior of alpha iron or at least of non-gamma iron. It is characterized, once the blue range is passed, by continuous decrease of strength and of stiffness and increased ductility. At 900 degrees Cent. (1650 degrees Fahr.) the alpha iron transforms into gamma iron, and this is accompanied by increased strength and stiffness and decreased ductility. We infer from this that gamma iron at the lowest temperature at which it can exist, that is, at 900 degrees Cent. plus is stronger and less ductile than alpha iron at the highest point at which it can exist, namely, 900 degrees Cent. minus. On further increase of temperature the strength and stiffness of gamma iron decrease while the ductility increases sharply, the metal undergoing 16 full twists at 1100 degrees Cent. (2010 degrees Fahr.) before rupture occurs. It is noted that ingot iron is more ductile at 500

degrees Cent. (930 degrees Fahr.) than it is slightly above 900 degrees Cent. (1650 degrees Fahr.) The fact that ingot iron suffers from red shortness between 900 and 1000 degrees Cent. appears to be strikingly related to the low ductility of gamma iron in this range of temperature.

In regard to the peak occurring in the ductility curve, at 800 degrees Cent. (1470 degrees Fahr.), I should like to whisper, if I dared, that it might correspond to the A_2 point, and that the sudden fall of ductility at that point might be due to the formation of beta iron, leaving to those who are so certain that a difference in magnetic properties is the only one existing between alpha and beta iron to give us a better explanation.

Carbon Steel (0.12 Per Cent). The three curves presented in Fig. 11 are in agreement in placing the blue heat phenomenon at or near 300 degrees Cent. (570 degrees Fahr.); but the peak in the factor of stiffness is very much more pronounced. There is a sharp decrease of strength and stiffness and a sharp increase of ductility once the blue heat range is passed. Discontinuities in the curves are observed in the vicinity of 700 degrees Cent. (1290 degrees Fahr.) and at 900 degrees Cent. (1650 degrees Fahr.), that is, when the thermal points A_1 and A_3 may be expected to occur. Confining our attention to the ductility curve it seems to indicate that the ductility of the aggregate of pearlite and alpha ferrite increases very rapidly with ascending temperatures up to 700 degrees Cent. when 15 full twists are required to break the bar. According to general belief the pearlite now transforms to austenite, and this is accompanied by a sudden fall of ductility, at 800 degrees Cent. (1470 degrees Fahr.), but 3½ twists being required to cause rupture. On further increase of temperature the aggregate of austenite and alpha ferrite believed to exist between A_1 and A_{3-2} increases rapidly in ductility reaching a second maximum (11 twists) at 900 degrees Cent. (1650 degrees Fahr.). This is followed by a sharp drop yielding a second minimum (6 twists) at 950 degrees Cent. (1740 degrees Fahr.). If as it is generally held the A_{3-2} point merely indicates on heating the *end* of ferrite absorption by the existing austenite, or on cooling the *beginning* of ferrite rejection, it is difficult to explain why it would be accompanied by so marked and sudden a change in ductility. It suggests as well as other considerations a possible wrong conception of the meaning of the point A_{3-2} and, when carbon is present, of the

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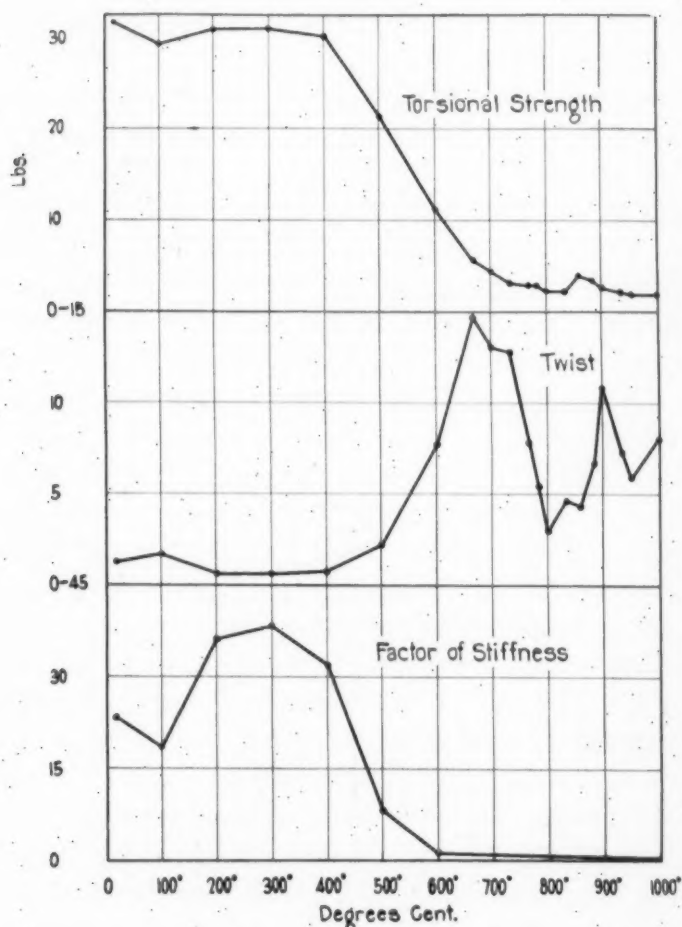metal under-
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luctile at 500

Fig. 11—Twisting Tests of 0.12 Per Cent Carbon Steel.

point A_3. Time, however, does not permit further discussion of this matter.

We would expect this 0.12 per cent carbon steel if at all red short to exhibit this tendency most markedly in the vicinity of 800 degrees Cent. (1470 degrees Fahr.) when the ductility of the austenite present is at its minimum. It should be noted, however, that it is now more ductile (3½ twists) than ingot iron at its minimum ductility (slightly more than two twists at 950 degrees Cent. (1740 degrees Fahr.). The mild steel under discussion is considerably less ductile at 800 (1470) than at 600 degrees Cent. (1110 degrees Fahr.).

Carbon Steel (0.38 Per Cent). The torsional strength curve shown in Fig. 12 places the blue heat phenomenon at 400 degrees
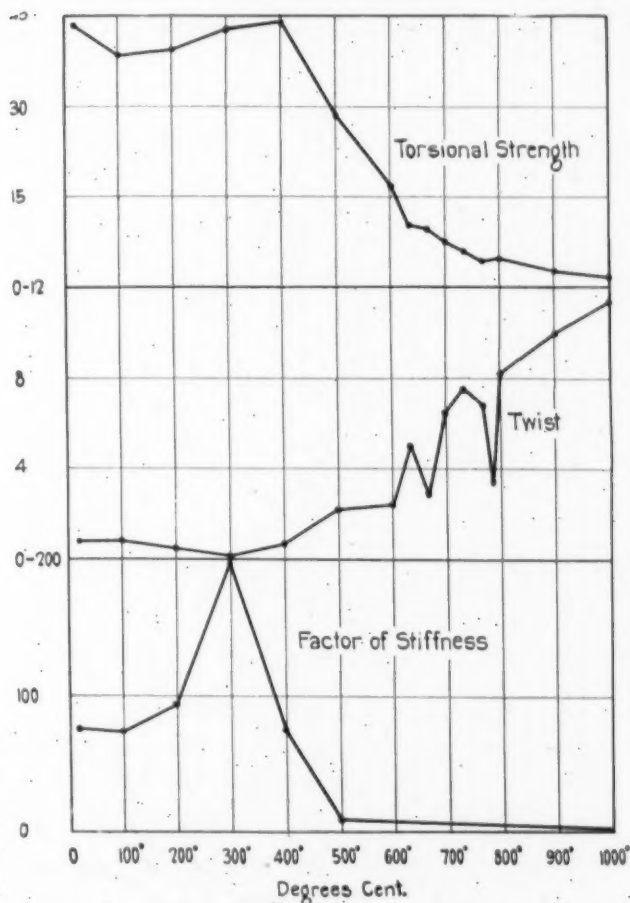
Fig. 12—Twisting Tests of 0.38 Per Cent Carbon Steel.

Cent. (750 degrees Fahr.), but the ductility and factor of stiffness curves, presented in Fig. 12, locate it at 300 degrees Cent. Attention is again called to the pronounced peak in the latter curve. After the blue heat range has been passed the decrease of strength and of stiffness and increase of ductility are very steep. Discontinuities occur in the ductility and factor of stiffness curves in the vicinity of the temperatures corresponding to the A_1 and A_{3-2} points. Only one discontinuity near the A_{3-2} point is observed in the torsional strength curve.

Referring to the ductility curve and comparing it with the corresponding curve of the 0.12 per cent carbon steel it is noted that the peaks are now closer together which corresponds to the closing up of the thermal points with increased carbon. We shall expect these peaks to merge into a single one in eutectoid steel.
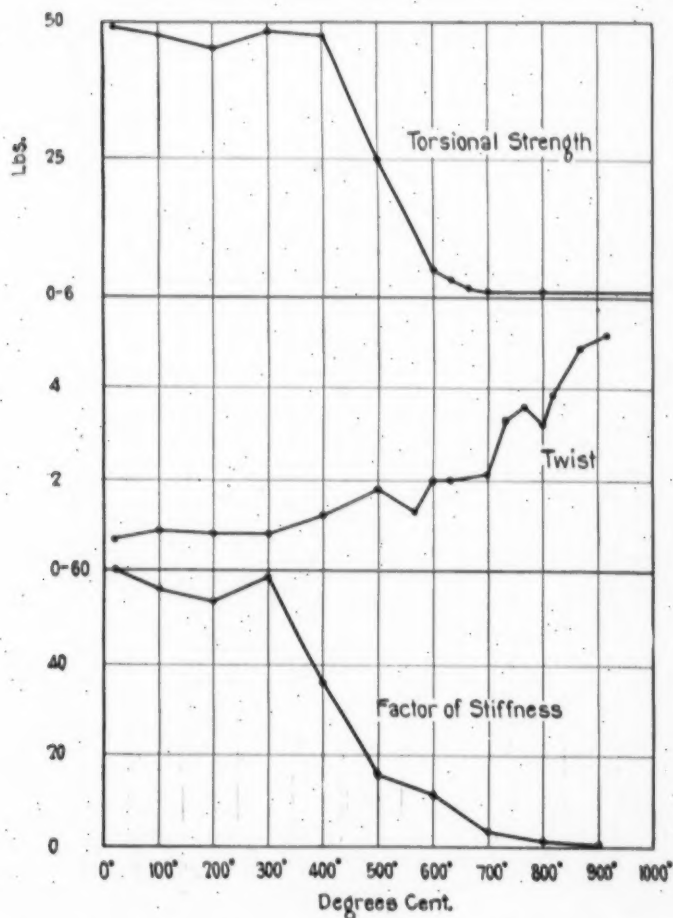
Fig. 13—Twisting Tests of 0.66 Per Cent Carbon Steel.

Here again we note increased ductility with increasing temperature in the alpha range, where the steel consists of an aggregate of pearlite and of alpha ferrite. At some 675 degrees Cent. (1245 degrees Fahr.) the pearlite is believed to be transformed into austenite and a sharp fall of ductility is noted reaching a minimum of 2¾ twists. Ductility is soon recovered, however, to attain a second maximum at 775 degrees Cent. (1425 degrees Fahr.) when it requires a little more than 7 twists to cause fracture. After passing the A_{3-2} point a second abrupt fall of ductility is observed to a minimum of 3½ twists. Further increase of temperature results in a very great increase of the ductility of the austenite now forming the totality of the steel. At 1000 degrees Cent. (1830 degrees Fahr.) the steel twists 11½ times. It will be noted in comparing the ductility curves of the 0.12 and 0.38 per cent carbon steels that in the former
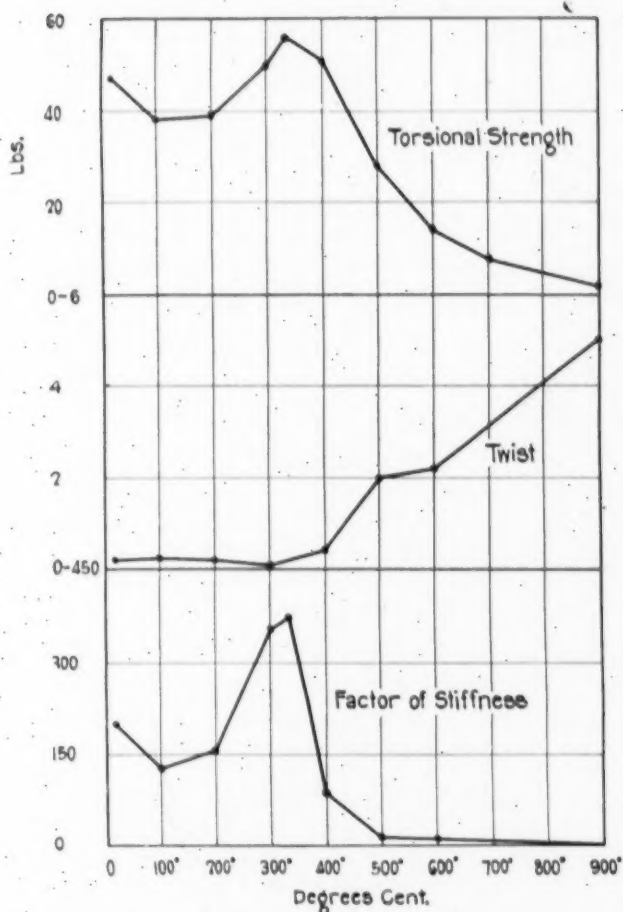
Fig. 14—Twisting Tests of 0.82 Per Cent Carbon Steel.

the first maximum exceeds the second maximum, while in the latter the second maximum exceeds the first. If our interpretation of these diagrams is correct it would follow that in the 0.12 per cent carbon steel the ferrite-pearlite aggregate immediately before its transformation is more ductile than after complete transformation, whereas in the 0.38 per cent carbon steel the reverse is true. Red shortness if it occurs at all in the 0.38 per cent carbon steel would be maximum in the vicinity of 700 degrees Cent. (1290 degrees Fahr.).

Carbon Steel (0.66 Per Cent.). The usual features are observed, and are shown in Fig. 13, i.e., maximum blue heat brittleness at 300 degrees Cent. (570 degrees Fahr.), and discontinuities in all curves in the vicinity of the thermal critical point or points. The fact that the ductility curve exhibits but one maximum and one minimum may be ascribed to the closeness of the points A_{3-2} and A_1 in a steel contain-
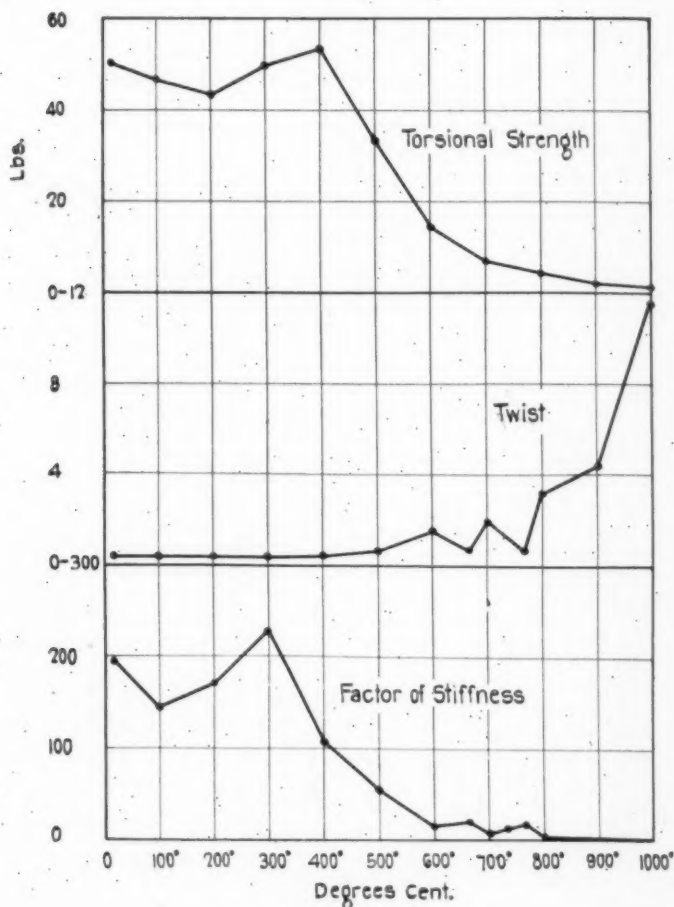
Fig. 15—Twisting Tests of 1.17 Per Cent Carbon Steel.

ing so much carbon. Our interpretation of these discontinuities remains the same and need not be repeated.

Carbon Steel (0.82 Per Cent.). The three curves as shown in Fig. 14 reveal the existence of the blue heat range at or near 300 degrees Cent. (570 degrees Fahr.). There is a pronounced peak in the factor of stiffness curve at 325 degrees Cent. (615 degrees Fahr.). The discontinuity in the curves expected to occur at about 700 degrees Cent. (1290 degrees Fahr.) is not apparent. It was obtained, however, in other eutectoid steels.

Carbon Steel (1.17 Per Cent.). The ductility curve as shown in Fig. 15 does not reveal blue heat brittleness while the torsional strength curve locates it at 400 degrees Cent. (750 degrees Fahr.) and the factor of stiffness curve sharply at 300 degrees Cent. (570

degrees Fahr.).   Discontinuities are observed in the vicinity of 700 degrees Cent. (1290 degrees Fahr.), that is, near the A_{3-2} point.

A fully satisfactory explanation of the blue heat phenomenon is still wanting.   The most recent attempt at supplying it was made by R. S. Dean, R. O. Day and J. L. Gregg,[1] who attribute blue heat brittleness to the presence of nitrogen.   On quenching, the nitride according to them is retained in solution causing a marked increase of hardness.   On aging at room temperature a further increase of hardness is observed, resulting from nitride dispersion.   On heating between 100 and 300 degrees Cent. (210 and 570 degrees Fahr.) the hardness is lowered because of the agglomeration of the nitride, while at 300 degrees Cent. (570 degrees Fahr.) the nitride again goes into solution resulting anew in increased hardness.   Without attempting a critical discussion of the theory, it may be pointed out that a recent carbon steel of German origin, know as Izett, is apparently completely free from blue heat brittleness, although it contains 0.0108 per cent nitrogen, while a steel of American manufacture of nearly the same chemical composition has a marked blue heat brittleness, although it contains less nitrogen, namely, 0.0086 per cent.   The analysis of these steels was performed in the laboratory of the Central Alloy Steel Corporation.

Torsional Strength of Armco Iron and of Carbon Steel.   It is noted (Fig. 16) that while at room temperature the strength varies greatly, from less than 30 pounds for the 0.12 per cent carbon steel to over 50 pounds for the 1.17 per cent carbon steel, and while at 400 degrees Cent. (750 degrees Fahr.) the difference is greater still, at higher temperatures a closing up of the curves occurs, and at 900 degrees Cent. (1650 degrees Fahr.) all the steels have the same strength, namely, about 3 pounds.   The actual figures are 0.12 per cent carbon, 3.05, 0.38 per cent carbon, 2.95, 0.66 per cent carbon, 3.30, 0.82 per cent carbon, 3.10 and 1.17 per cent carbon, 3.02 pounds. It is interesting to note that all the steels are now in their austenitic range.   The torsional strength of ingot iron at 900 degrees Cent. (1650 degrees Fahr.) is less than that of the steels, namely, 1.9 pounds, but it should be considered that it may not at this temperature be fully in its gamma condition,—at temperatures slightly above 900 degrees Cent. (1650 degrees Fahr.) its strength increases.   The in-

[1] R. S. Dean, R. O. Day and J. L. Gregg, "Relation of Nitrogen to Blue Heat Phenomena in Iron and Dispersion Hardening in the System Iron-Nitrogen." American Institute of Mining and Metallurgical Engineers, February, 1929.

teresting conclusion appears to be justified that in the range in which it is stable the strength of gamma iron is not affected by the percentage of carbon it may contain: 1.17 per cent carbon does not produce stronger austenite than 0.12 per cent carbon.

Twist of Armco Iron and of Carbon Steels. While as it has been

Fig. 16—Torsional Strength of Ingot Iron and of Some Carbon Steels Between Room Temperature and 900 Degrees Cent. (1650 Degrees Fahr.).

just noted all steels have the same strength at 900 degrees Cent. (1650 degrees Fahr.) their ductility at that temperature varies within wide limits. (Fig. 17). In order of decreased ductility they are classified as follows: 0.38 per cent carbon, 0.12 per cent carbon, 0.82 and 0.66 per cent carbon, 1.17 per cent carbon and Armco iron. The low ductility of Armco iron is accounted for when it is considered that it is now in its red short range as previously explained. The various steels do not maintain the positions they occupy at room temperature. Indeed the 0.38 per cent carbon steel which is the fourth in ductility at room temperature has the greatest ductility at 900 degrees Cent. (1650 degrees Fahr.). The 0.12 per cent carbon steel acquires great ductility at 700 degrees Cent. (1290 degrees Fahr.), but loses it at 800 degrees Cent. (1470 degrees Fahr.). Additional experiments and closer study will be required before any attempt can be made at interpreting these results.

Factor of Stiffness of Armco Iron and of Carbon Steels. The blue heat phenomenon is shown to occur at 300 degrees Cent. (570

degrees Fahr.) in every case, being most pronounced in the steels containing respectively 0.82, 1.17, and 0.38 per cent carbon (Fig. 18). The position of the factor of stiffness of the 0.66 per cent carbon steel seems abnormal and is difficult to explain. At room temperature the 0.82 and 1.17 per cent carbon steels have the same stiffness, 0.38 and 0.66 per cent steels nearly the same, and 0.12 per cent carbon
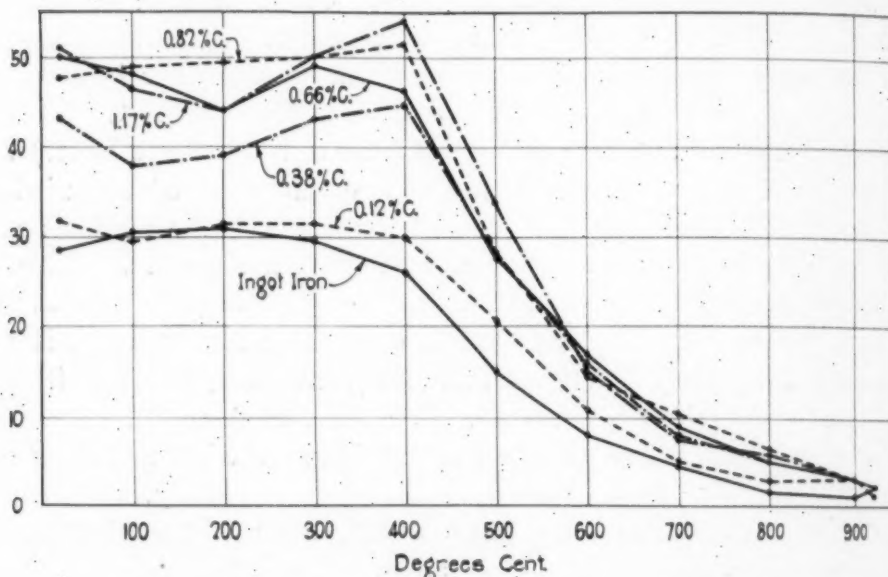

Fig. 17—Twist of Ingot Iron and of Some Carbon Steels Between Room Temperature and 900 Degrees Cent. (1650 Degrees Fahr.).

steel and Armco iron also about the same. At 500 degrees Cent. (930 degrees Fahr.) the factors of stiffness of all samples have decreased sharply while at temperatures exceeding 700 degrees Cent. (1290 degrees Fahr.) they are so small that they cannot be separated on the scale of the diagram.

Twist of Some Eutectoid and Hypereutectoid Steels. Extraordinary ductility suddenly acquired by some eutectoid and hypereutectoid steels at temperatures exceeding some 700 to 800 degrees Cent. (1290 to 1470 degrees Fahr.) is illustrated in Fig. 19. A steel containing 0.88 per cent carbon twisted 1.75 revolutions at 700 degrees Cent. (1290 degrees Fahr.), 27.75 revolutions at 800 degrees Cent. (1470 degrees Fahr.), 63 at 900 degrees Cent. (1650 degress Fahr.) and 70 at 1000 degrees Cent. (1830 degrees Fahr.).
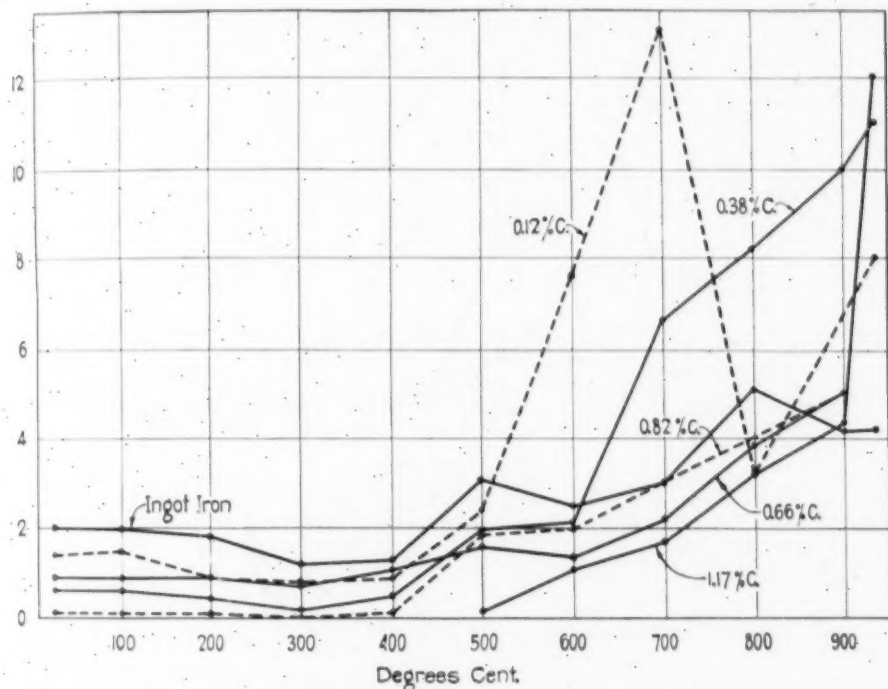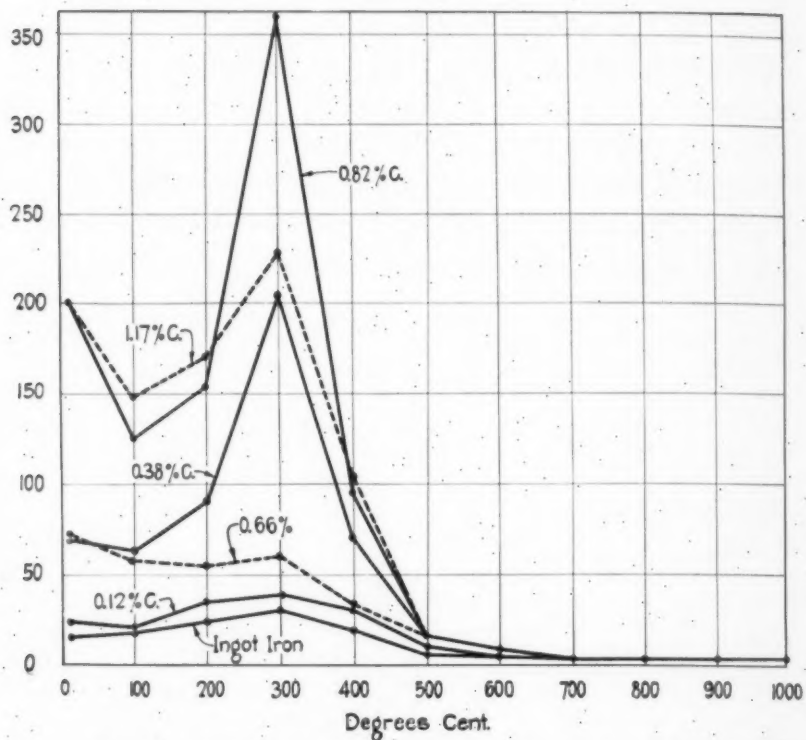
Fig. 18—Factor of Stiffness of Ingot Iron and of Some Carbon Steels Between Room Temperature and 900 Degrees Cent. (1650 Degrees Fahr.).
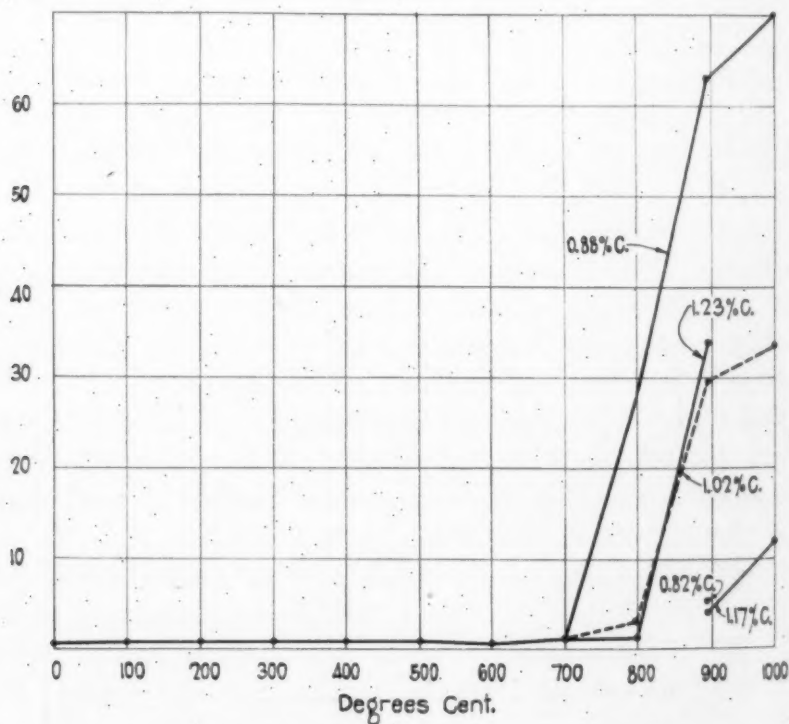

Fig. 19—Twist of Some Eutectoid and Hypereutectoid Steels Between Room Temperature and 1000 Degrees Cent. (1830 Degrees Fahr.).

Another steel containing 1.02 per cent carbon twisted 3.07 revolutions at 800 degrees Cent. (1470 degrees Fahr.), 29.5 revolutions at 900 degrees Cent. (1650 degrees Fahr.) and 33.75 at 1000 degrees Cent. (1830 degrees Fahr.). Another steel containing 1.23 per cent carbon twisted 1.78 revolutions at 800 degrees Cent. (1470 degrees Fahr.) and 34 revolutions at 900 degrees Cent. (1650 degrees Fahr.). On the other hand another eutectoid steel containing 0.82 per cent carbon twisted but 5 revolutions at 900 degrees Cent. (1650 degrees Fahr.) and another steel containing 1.17 per cent carbon twisted only 4.28 revolutions at 900 degrees Cent. (1650 degrees Fahr.) and 12 revolutions at 1000 degrees Cent. (1830 degrees Fahr.). No explanation can be offered at this time to account for such marked difference in the behavior of some steels of apparently like composition.

It may be advisable to point out that this sudden increase of ductility is acquired in the vicinity of the thermal critical points. It is believed that the steels are then in their austenitic range.

No hypoeutectoid steel so far tested has exhibited this phenomenon.

An attempt at bringing out the relations which appear to exist between the critical twists and the thermal critical points is offered in Fig. 20. The ductility of alpha iron represented by Armco iron increases from room temperature to the A_3 point when it is transformed into gamma iron. This corresponds to a fall of ductility duly represented in the twisting curve and indicates that gamma iron when first formed is considerably less ductile than alpha iron immediately before its transformation. With increasing temperature the ductility of gamma iron increases rapidly. In the presence of 0.12 per cent carbon the ductility of the ferrite-pearlite aggregate increases from room temperature to the A_1 point when a sharp fall of ductility is recorded. The ductility of the ferrite-austenite aggregate existing between A_1 and A_3 increases until the A_3 point is reached when there is another sharp fall of ductility. The steel is now entirely austenitic and its ductility increases rapidly with further increase of temperature. With 0.38 per cent carbon similar critical twists are observed, but within a narrower range of temperature as the A_1 and A_{3-2} points are now closer together. When the steel is of eutectoid composition only one critical twist is observed, generally of small magnitude. While we would expect it to occur at the point Ar_{3-2}, when pearlite changes to austenite, the results of our experiments place it

nearer 600 degrees Cent. (1110 degrees Fahr.). Further tests may change its position.

Red Shortness. These results indicate that red shortness in iron and steel when it exists is most intense slightly above the A_3
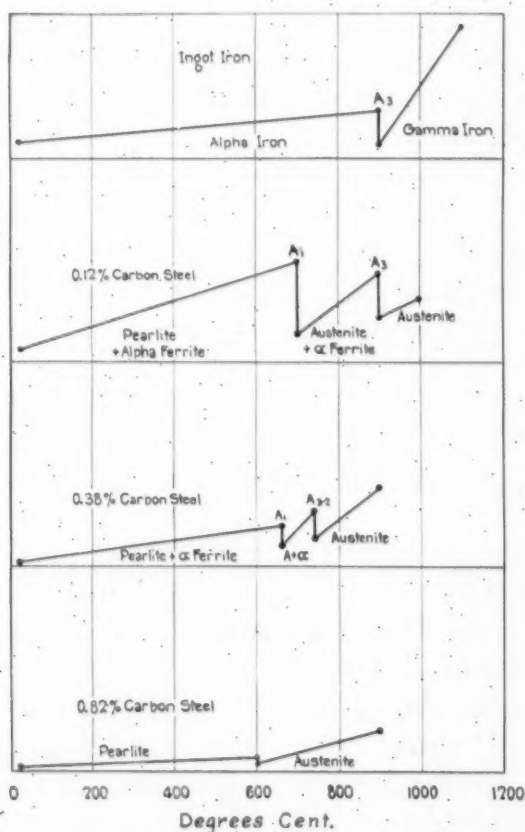
Fig. 20—Relation Between Critical Twists and Thermal Critical Point of Ingot Iron and of Some Carbon Steels.

point when gamma iron is about to transform into alpha iron on cooling, or when it has just come into existence on heating. It is believed that alpha iron is never red short. To put it in another way, red shortness appears to be a property of face-centered (gamma) iron and to be at its maximum intensity at the lowest temperature at which gamma iron can exist. These remarks, of course, apply only to carbon steels.

Austenitic Steels. Two austenitic steels were subjected to our twisting tests. (Fig. 21.) They had the following composition.
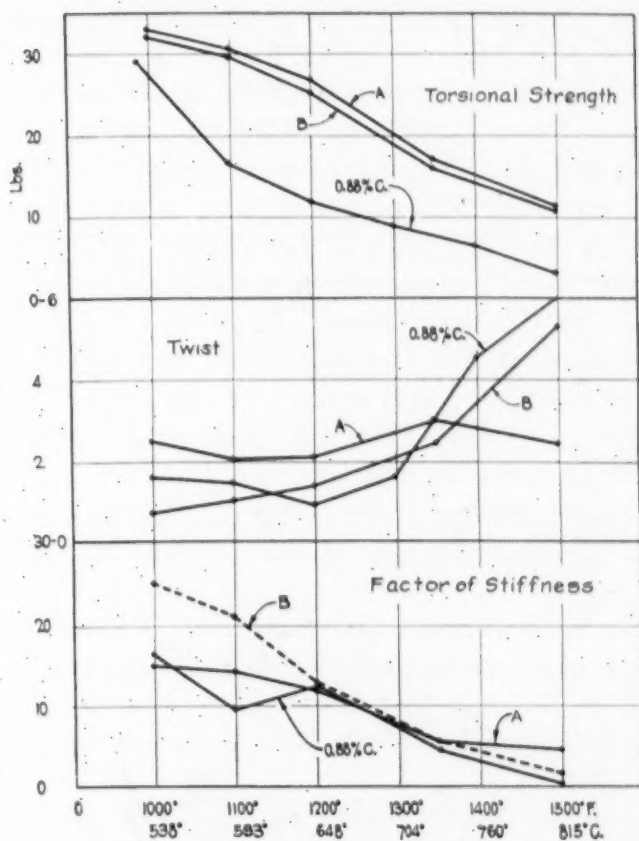
Fig. 21—Torsional Tests of Nickel-Chromium Austenitic Steels and a Eutectoid Carbon Steel.

Designation	C	Si	Mn	P	S	Ni	Cr
A	0.08	0.36	0.35	0.022	0.014	8.61	17.20
B	0.18	1.24	0.59	0.022	0.019	19.83	7.66

It will be noted that one is substantially the reverse of the other so far as the contents of nickel and chromium are concerned. These steels were twisted at the following temperatures, 540 degrees Cent. (1000 degrees Fahr.), 595 degrees Cent. (1100 degrees Fahr.), 650 degrees Cent. (1200 degrees Fahr.), 735 degrees Cent. (1350 degrees Fahr.) and 815 degrees Cent. (1500 degrees Fahr.). The graph also includes a carbon steel containing 0.88 per cent carbon. It will be noted that the torsional strength of both austenitic steels shows a steady decrease and that steel A remains throughout slightly stronger than steel B. The austenitic steels are at all temperatures considerably stronger than the carbon steel, their superiority in this regard being greatest at 650 degrees Cent. (1200 degrees Fahr.). At 540

degrees Cent. (1000 degrees Fahr.) steel A is more ductile and steel B less ductile than the carbon steel, but at 650 degrees Cent. (1200 degrees Fahr.) they are both more ductile than the carbon steel. Up to 735 degrees Cent. (1350 degrees Fahr.) steel A is more ductile than steel B, but at 815 degrees Cent. (1500 degrees Fahr.) steel B is much more ductile. This reversal is apparently due to a loss of ductility which steel A suffers in the vicinity of 735 degrees Cent. (1350 degrees Fahr.), a phenomenon which has been observed by others in steel of this composition. B. Strauss found a marked fall of tensile ductility at 700 degrees Cent. (1290 degrees Fahr.) in a steel containing 0.17 per cent carbon, 20 per cent chromium and 7 per cent nickel. Dr. H. H. Lester in a steel containing 0.095 per cent carbon, 18.20 per cent chromium, and 8.84 per cent nickel noted a similar fall at 760 degrees Cent. (1400 degrees Fahr.), and believes that the real minimum is close to 719 degrees Cent. (1325 degrees Fahr.).

From the results which we have obtained, it would seem as if the steel B type of nickel-chromium austenitic steel does not undergo this loss of ductility, although the presence of 1.24 per cent silicon in this steel may be responsible rather than the difference of nickel and chromium contents.

It is significant that Dr. Lester reports the occurrence of a sharp thermal critical point in the type of steel represented by steel A in cooling at 617 degrees Cent. (1140 degrees Fahr.) and on heating at 716 degrees Cent. (1320 degrees Fahr.), while in a similar determination, which he kindly conducted at my request, no thermal point was detected in the type of steel represented by steel B.

At 650 degrees Cent. (1200 degrees Fahr.) and at 735 degrees Cent. (1350 degrees Fahr.) the three steels have nearly the same factors of stiffness, while at 815 degrees Cent. (1500 degrees Fahr.) the austenitic steels are stiffer, steel A being now superior in this respect to steel B.

Non-Austenitic Chromium Steels

The twisting test (Fig. 22) was applied to three brands of non-austenitic chromium steels of the following compositions

Designation	C	Si	Mn	P	S	Cr	Ni
C	0.07	1.17	0.34	0.018	0.010	17.71	0.22
D	0.31	0.32	0.68	0.035	0.013	26.67	0.37
E	0.17	2.88	0.22	0013	0.011	3.26	0.18

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Cr	Ni
17.71	0.22
26.67	0.37
3.26	0.18

Fig. 22—Torsional Tests of Non-austenitic Chromium Steels.

at 540 degrees Cent. (1000 degrees Fahr.), 595 degrees Cent. (1100 degrees Fahr.), 650 degrees Cent. (1200 degrees Fahr.), 735 degrees Cent. (1350 degrees Fahr.) and 815 degrees Cent. (1500 degrees Fahr.). It will be seen that the three steels do not differ much in torsional strength at any of the temperatures at which they were tested, steel C being, except at 650 degrees Cent. (1200 degrees Fahr.), the strongest and steel E the least strong. While they do not differ greatly in ductility at room temperature, steel C at 735 degrees Cent. (1350 degrees Fahr.) is much more ductile than the other two, 43.85 twists being required to cause its fracture. On further increase of temperature its ductility, although remaining very high, is substantially lowered. No such discontinuities are observed in the ductility curve of steel D and steel E, which, nevertheless, remains at 735 degrees Cent. (1350 degrees Fahr.) and at 815 degrees Cent (1500 degrees Fahr.), considerably less ductile than steel C. Steel E increases in ductility with increasing temperature much less than

Fig. 23—Torsional Tests of Two Chromium
Steels and 0.88 Per Cent Carbon Steel.

steel C and steel D. While the factor of stiffness of steel E at room
temperature is much lower than the factor of stiffness of steel D
and steel C at 650 degrees Cent. (1200 degrees Fahr.) the three
steels have practically the same stiffness. At higher temperatures the
factors are too small to be indicated on the scale of the diagram.

Steel A, Steel C and 0.88 Per Cent Carbon Steel. The char-
acteristics, as revealed by torsional tests, of an austenitic nickel-
chromium steel of steel A type, of a non-austenitic chromium steel
of the steel C type, and of a carbon eutectoid steel (0.88 per cent
carbon) have been assembled in Fig. 23. Steel C was tested only at
temperatures between 540 and 815 degrees Cent. (1000 and 1500 de-
grees Fahr.). While at room temperature the strength of the carbon
steel is considerably less than the strength of steel A, between 200
and 400 degrees Cent. (390 and 750 degrees Fahr.) it has greater
strength, this being due to the blue heat phenomenon. At all higher

temperatures the strength of steel A remains greater. The strength of steel C at the temperatures at which it was tested is about half way between.

The three steels have relatively low ductility, at temperatures below some 600 degrees Cent. (1120 degrees Fahr.). The ductility of steel A remains low, never exceeding 2.5 revolutions and being minimum at 700 degrees Cent. (1290 degrees Fahr.), although this cannot be indicated on the scale of the diagram. In the vicinity of 700 degrees Cent. steel C acquires suddenly the extraordinary ductility already noted, and the eutectoid steel at a somewhat higher temperature a still greater ductility.

The factor of stiffness curves bring out clearly the blue heat brittleness of the carbon steel and the complete absence of a corresponding range in the austenitic steel. Another set of experiments has revealed the presence of a blue heat range in steel C.

TIME-STRAIN DIAGRAMS

In order to secure, if possible, additional information of the behavior of steel when under plastic deformation, a revolving drum as already explained was so connected with the twisting pulley that autographic time-strain curves could be obtained. While the present author originated these twisting tests and devised the necessary apparatus, the addition of the recording attachment is due to R. E. Leiter. Unexpected results were obtained, and it became apparent that each kind of steel has its own sharply defined characteristics under plastic deformation at room and at elevated temperatures. The close agreement between diagrams obtained from duplicate tests eliminate the possibility of these characteristics being due to variations in the speed of loading, to personal equation or to other variables. It is believed that if we could interpret correctly their significance, a deeper knowledge of the nature of steel would be obtained. Time and space permit only the consideration of a few of these diagrams. To clarify let us consider two characteristic diagrams; the smooth type and the step-like or jerky type, as illustrated in Fig. 24. In the smooth type the test started at S. At A the bar had twisted 180 degrees, the time required being two minutes (each vertical inch represents thirty seconds). At B it had twisted one full revolution, at M one and one-half revolutions, at D two revolutions, and it finally broke at R. The point M at the bottom of the diagram

corresponds to the point M' at the top when the paper is wrapped
around the drum. The total twist was two revolutions and 135 de-
grees or 855 degrees and the total time sixteen (the number of ver-
tical inches travelled) times thirty seconds; that is, eight minutes or
480 seconds. The average time per revolution was, therefore,
480×360/855 or 3 minutes and 22 seconds. The first full
twist required four minutes and the second three minutes. The
difference, namely one minute, may be designated as twisting
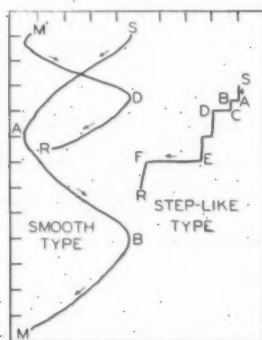
Fig. 24—Smooth and
Step-like Types of
Time-Strain Diagrams.
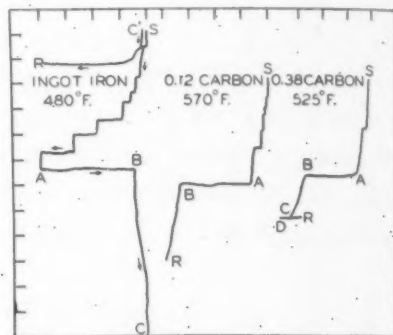
Fig. 25—Time-Strain Diagrams of
Ingot Iron and Hypoeutectoid Steels
in the Blue Heat Range.

acceleration. It has been found that, although different steels may
rupture under the same load, the number of twists may not only vary
but also the average twisting speed and the twisting acceleration. In
the step-like type the test started at S. From S to A there was no
twist. At A a sudden twist from A to B of some 13 degrees. From
C to D, covering a period of 21 seconds, there was no twist. At
D another sudden twist, and so on. At E a sudden twist of over
90 degrees and finally fracture at R, the total twist being somewhat
less than 180 degrees. It will be shown that this type of diagram is
characteristic of the plastic deformation of ferrite in the blue heat
range.

It should be borne in mind that these time-strain diagrams
correspond to a predetermined and constant rate of twist, namely,
about 10 degrees in 10 seconds, at least until the metal takes the
matter into its own hands and twists at a faster speed without further
increase of load. With a more rapid twisting speed these curves
might be deeply altered, probably through shorter periods of rest
between steps. The chief interest, for the moment, of these two
different types of time-strain diagrams lies in the light it throws on

the different behavior of steel when subjected to plastic deformation in the blue heat range and outside of that range.

It seems appropriate to recall the step-like stress-strain diagrams obtained by A. V. de Forest[2] in testing by tension some annealed phosphor bronze wire and some annealed brass wire. They resulted from increasing the temperature of testing to 183 degrees Cent., while maintaining the speed of loading constant and also by increasing the speed of loading from some two minutes to twenty-seven minutes while maintaining the temperature constant. Mr. de Forest explains the phenomenon in the light of the assumed existence and behavior of amorphous cement. A smooth curve would indicate a relatively slow "setting" of the cement. Increased temperature or decreased speed of loading would result in a less viscous cement or in a longer time for the amorphous cement to yield before the load is again sufficient to produce steps, in either case yielding the step-like form of stress-strain diagrams.
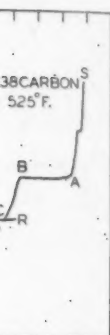
ARMCO IRON AND HYPOEUTECTOID STEEL IN THE BLUE HEAT RANGE

The plastic deformation of ingot iron in the blue heat range takes place by a series of sudden slips followed by short periods of rest. (Fig. 25.) Starting the test at S the bar had twisted 180 degrees at A after two and one-half minutes, through a number of slips showing a tendency to increase in magnitude. At A the metal after a period of rest of some thirty seconds twisted suddenly some 157 degrees. From B to CC' the twisting was slight and gradual and the iron finally ruptured at R after having undergone another abrupt twist. The total twist was about one and one-half revolutions. This time-strain diagram throws considerable light on the behavior of alpha iron under stress in the blue heat range.

Steel containing 0.12 per cent carbon and, therefore, a considerable amount of free alpha iron undergoes torsional deformation in the blue heat range in a similar manner; that is, through a number of successive yieldings. We also note that a large twist AB of some 120 degrees occurred immediately before rupture took place at R. Compared with the behavior of ingot iron, this mild steel does not undergo as many distinct twists, and the total angle is but 180 degrees.

[2]A. V. de Forest, "Some Experiments on the Plastic Elongation of Wire," *Proceedings* American Society for Testing Materials, Vol. XVI, 1916, p. 455.

With 0.38 per cent carbon and therefore less free alpha iron the number of sharp twists is reduced to two, and we again note the occurrence of a pronounced slip AB of about 103 degrees. At C the bar twisted to D and then back to R, where it failed. The amount
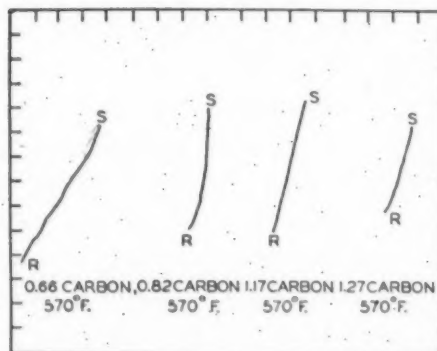
Fig. 26—Time-Strain Diagrams of Eutectoid and Hypereutectoid Steels in the Blue Heat Range.
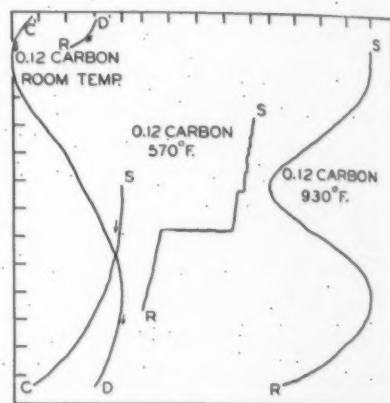
Fig. 27—Time-Strain Diagrams of 0.12 Per Cent Carbon Steel, Below, At and Above the Blue Heat Range.

of twist immediately before fracture is, therefore, CD + DR, or about 57 degrees.

EUTECTOID AND HYPEREUTECTOID STEEL IN THE BLUE HEAT RANGE

In the absence of free alpha iron, the time-strain diagrams of steels tested in the blue heat range consist of smooth curves, the step-like appearance so characteristic of the curves of iron and hypoeutectoid steels in this range being absent. (Fig. 26.) Even with 0.66 per cent carbon, the curve is free from jerks, although such steel contains some 30 per cent of ferrite. It seems, therefore, that the step-like mode of plastic deformation observed in the blue heat range of iron and of hypoeutectoid steels must be ascribed to the behavior of alpha iron. It does not follow, however, that eutectoid and hypoeutectoid steels do not undergo a loss of ductility and an increase in strength and hardness in the blue heat range, although the phenomenon is generally less pronounced, especially in regard to increased strength.

0.12 PER CENT CARBON STEEL BELOW, AT AND ABOVE THE BLUE HEAT RANGE

At room temperature the usual smooth curve is observed. (Fig.

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27.) The test started at S and rupture took place at R after a twist of one turn and 90 degrees.   At 300 degrees Cent. (570 degrees Fahr.) the step-like curve characteristic of the blue heat range is ob-
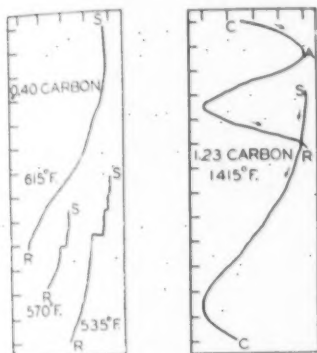
Fig. 28—Time-Strain Diagrams of 0.40 Per Cent Carbon Steel At and Above the Blue Heat Range.   Fig. 29— Time-Strain Diagrams of 1.23 Per Cent Carbon Steel at 770 Degrees Cent. (1420 Degrees Fahr.).
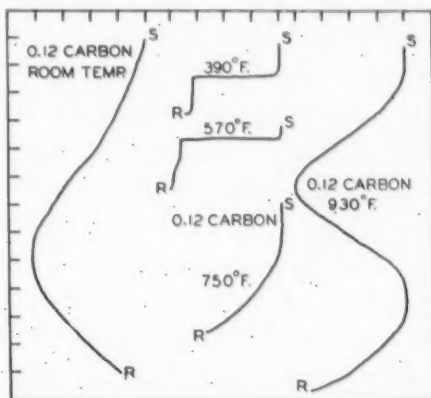
Fig. 30—Time-Strain Diagrams of 0.12 Per Cent Carbon Steel Twisted After Cold Working 25 Per Cent by Tension.

tained. At 500 degrees Cent. (930 degrees Fahr.) the curve is again smooth, the steel twisting nearly one and one-half revolutions.

0.40 Per Cent Carbon Steel at and Above the Blue Heat Range

At 280 and at 300 degrees Cent. typical step-like curves are obtained (Fig 28), while at 325 degrees Cent. the curve is smooth, from which it is inferred that the blue range of temperature has now been passed.

1.23 Per Cent Carbon Steel at 770 Degrees Cent.

The twisting started at S and the first twist was completed at A, requiring 330 seconds.   (Fig. 29.) The bar broke at R after two revolutions.   The second twist required 111 seconds.   What has been called the twisting acceleration was, therefore, 219 seconds and the average speed 220 seconds per twist.

Twisting After Cold Working

An attempt was made to study the influence of cold work on the twisting characteristics at various temperatures.   (Fig. 30.)   Refer-

ence will only be made here to some results obtained with steel containing 0.12 per cent carbon, the reduced sections of the test bars having been elongated 25 per cent in a tensile testing machine before applying the twisting test.   At room temperature a smooth curve is obtained, the bar twisting very slowly and finally rupturing after 360 seconds and a twist of some 315 degrees.   At 200 and 300 degrees Cent. (390 and 570 degrees Fahr.) the bars yielded suddenly and twisted some 135 degrees.   This was followed by a period of rest and another short twist immediately before fracture.   The marked alteration produced by cold work in the appearance of the curve in the blue heat range should be noted.   The many steps of the steel tested in an annealed condition have been replaced by a single one of considerable dimension.   This may be considered as an intensifying of the blue heat brittleness, and is in line with the view generally held and well supported by tensile and hardness tests.

At 400 degrees Cent. (750 degrees Fahr.) the curve is again smooth, the steel twisting some 130 degrees and the time required being 135 seconds.   At 500 degrees Cent. (930 degrees Fahr.) the curve is smooth and the twist about $1\frac{1}{2}$ revolutions.

TIME-STRAIN DIAGRAMS OF NICKEL-CHROMIUM AUSTENITIC STEELS

Steels of the 18 per cent chromium, 8 per cent nickel and 8 per cent chromium and 18 per cent nickel type were twisted at 540, 595, 650, 735 and 815 degrees Cent. (1000, 1100, 1200, 1350 and 1500 degrees Fahr.).

The time-strain diagrams of Fig. 31 exhibit the step-like mode of plastic deformation at 540, 595 and 650 degrees Cent. (1000, 1100 and 1200 degrees Fahr.).   At 735 degrees Cent. (1350 degrees Fahr.) and at 815 degrees Cent. (1500 degrees Fahr.) the curves are smooth. It is of interest to note that this mode of yielding under torsion which we have so far associated with the behavior of alpha iron in the blue heat range is apparently also a characteristic of gamma iron—or at least of austenite at temperatures of some 500 to 650 degrees Cent. (930 to 1200 degrees Fahr.).

The time-strain diagrams of Fig. 32 steel are similar to those of Fig. 31, although these two steels are of what may be termed "reversed" composition.   At 540, 595 and 650 degrees Cent. (1000, 1100 and 1200 degrees Fahr.) the curves are decidedly step-like in appear-

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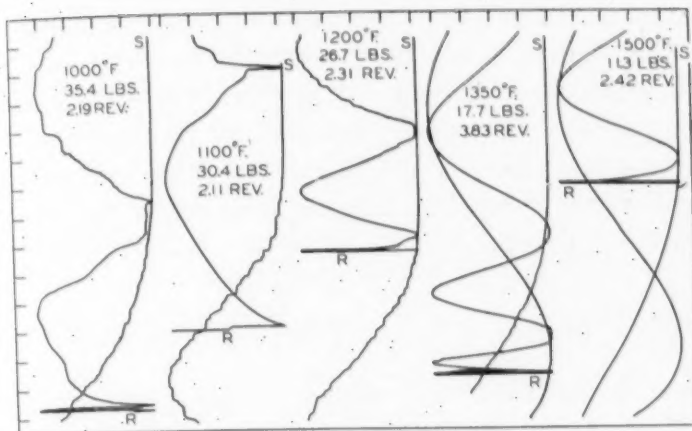t. (1000, 1100
like in appear-

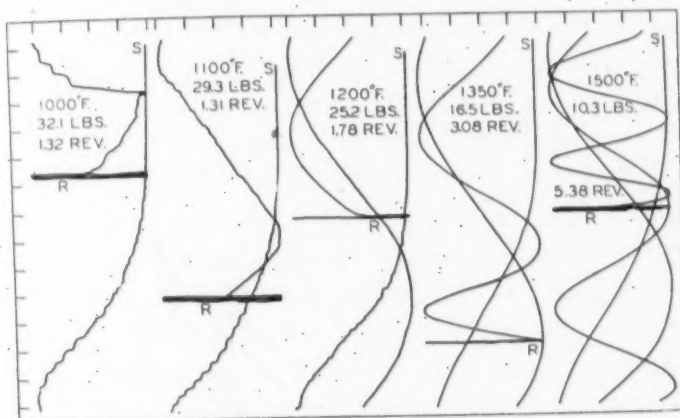Fig. 31—Time-Strain Diagrams of Nickel-Chromium Aus-
tenitic Steel. Steel A Type.

Fig. 32—Time-Strain Diagrams of Nickel-Chromium Aus-
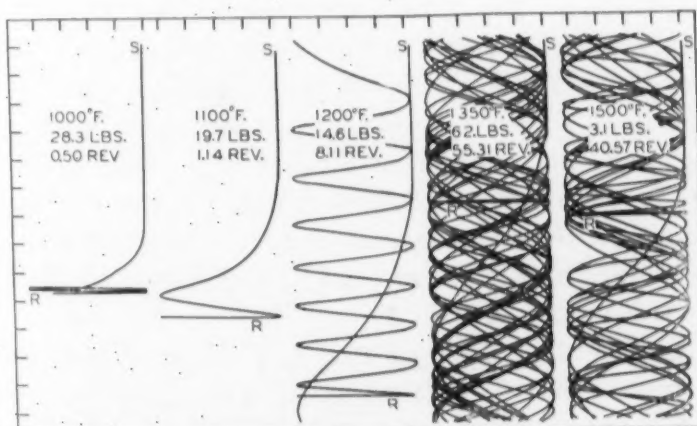tenitic Steel. Steel B Type.

Fig. 33—Time-Strain Diagrams of Non-austenitic Chromium
Steel. Steel C Type.

ance, while at 1350 and 1500 degrees Fahr. they are smooth. From the occurrence of these step-like curves in nickel-chromium austenitic steels one is led to consider the possibility of these steels having a blue heat range, although not readily detected by the usual method. At 815 degrees Cent. (1500 degrees Fahr.) the 8 per cent nickel 18 per cent chromium steel twists twice as much as the 18 per cent nickel 8 per cent chromium steel.

TIME-STRAIN DIAGRAMS OF NON-AUSTENITIC CHROMIUM STEELS

Three different chromium steels (steels C, D and E) were twisted at 540, 595, 650, 735 and 815 degrees Cent. (1000, 1100, 1200, 1350 and 1500 degrees Fahr.). The composition of these steels has been given on page 434. In the curves (Fig. 33) there is no indication of any step-like mode of yielding within the range of temperature covered. From 540 to 650 degrees Cent. (1000 to 1200 degrees Fahr.) there is a gradual increase of ductility. At 735 degrees Cent. (1350 degrees Fahr.), however, an extraordinary plasticity has been acquired, the bar requiring over 47 revolutions to cause its fracture. It recalls the ductility of some eutectoid and hypereutectoid steel in the vicinity of 700 degrees Cent. (1290 degrees Fahr.). It is believed that when steels exhibit such plasticity they are in their austenitic range. It would then be associated with the ductility of gamma or face-centered iron. At 815 degrees Cent. (1500 degrees Fahr.) the ductility of steel C is substantially less, 40.57 revolutions being required to break the bar. The diagram obtained at 650 degrees Cent. (1200 degrees Fahr.) affords a good example of increasing twisting speed. The 8 revolutions needed to cause rupture required, respectively, 330, 54, 48, 46, 42, 39, 39 and 33 seconds. The average twisting speed was, therefore, 79 seconds and the twisting acceleration 297 seconds.

The absence of step-like twisting is also observed in steel D. (Fig. 34.) It is more ductile than steel C at 540 degrees Cent. (1000 degrees Fahr.), and at 595 degrees Cent. (1100 degrees Fahr.), but it fails to acquire at 735 degrees Cent. (1350 degrees Fahr.) anything like the plasticity of steel C.

The increase of ductility of steel E (Fig. 35) is gradual from room temperature to 815 degrees Cent. (1500 degrees Fahr.) requiring respectively at the temperatures tested 0.38, 1.18, 5.89, 13.19 and
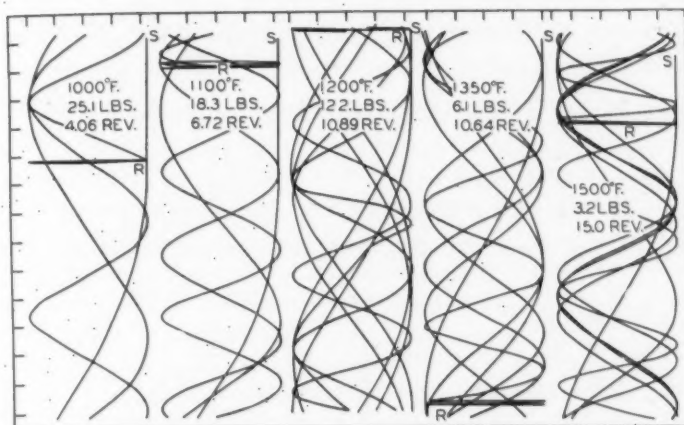
Fig. 34—Time-Strain Diagrams of Non-austenitic Chromium Steel. Steel D Type.
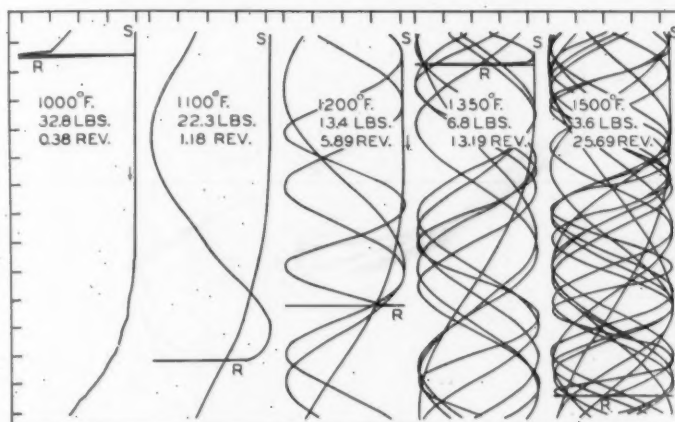
Fig. 35—Time-Strain Diagrams of Non-austenitic Chromium Steel. Steel E Type.
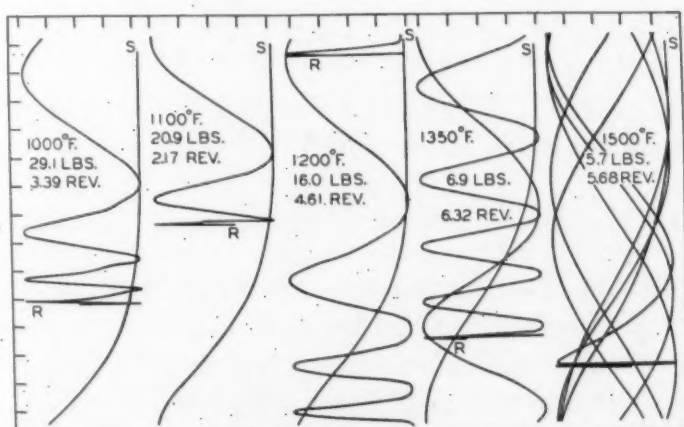
Fig. 36—Time-Strain Diagrams of S.A.E. 3130 Chromium-Nickel Steel.

25.69 revolutions. None of the curves exhibit any step-like appearance.

NICKEL-CHROMIUM STEELS S. A. E. 3130

This steel, having 0.35 per cent carbon, 1.25 per cent nickel, and 0.60 per cent chromium, was twisted at the five temperatures already referred to. There is no indication of step-like deformations in any of
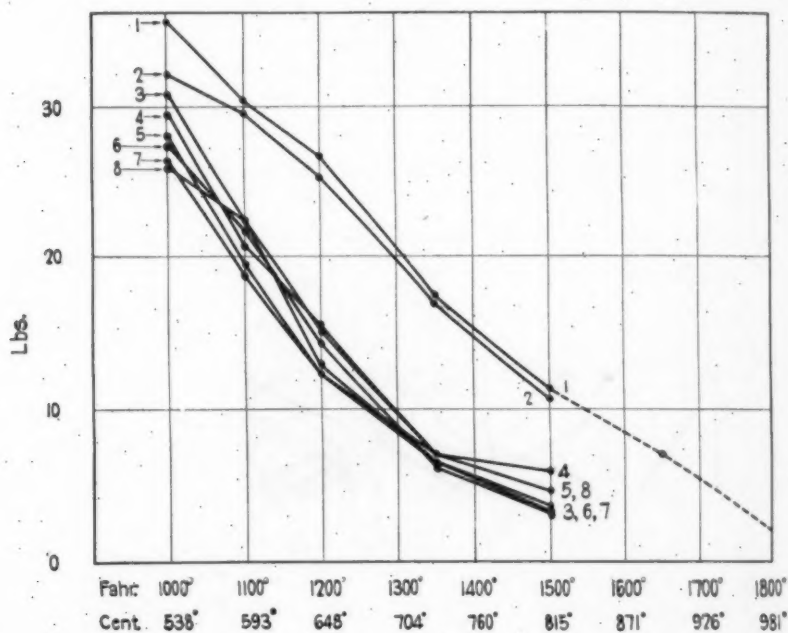
Fig. 37—Torsional Strength Versus Creep Stresses.

the curves. (Fig. 36.) It is less ductile at 595 degrees Cent. (1100 degrees Fahr.) than at 540 degrees Cent. (1000 degrees Fahr.). It is also less ductile at 735 degrees Cent. (1350 degrees Fahr.) than at 815 degrees Cent. (1500 degrees Fahr.) while the twisting speed at the latter temperature is much slower.

TORSIONAL STRENGTH VERSUS CREEP STRESSES

An extensive investigation was conducted at the Massachusetts Institute of Technology by Professor F. H. Norton, and some of the results obtained by him are shown graphically in Figs. 37 and 38. They indicate the creep stresses for a life of 100,000 hours with one per cent elongation at 1000, 1100, 1200, 1350 and 1500 degrees Fahr.

These tests included five nickel-chromium austenitic steels, five

non-austenitic chromium steels, one non-austenitic nickel-chromium steel, one nickel steel, one chromium-vanadium steel, two tungsten steels, one carbon steel and one low manganese steel. Seven of the steels used by Mr. Norton were subjected to our twisting tests at the temperatures employed by him. The steels were nickel-chromium
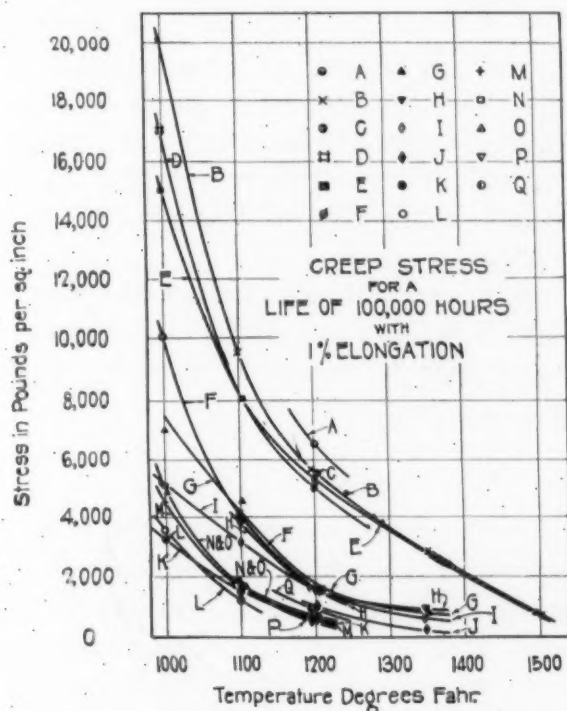
Fig. 38—Creep Stress for a Life of 100,000 Hours with 1 Per Cent Elongation.

austenitic steels 8 and 18 nickel-chromium and and 18 and 8 nickel-chromium, non-austenitic nickel-chromium steels, C, D, and E, S.A.E. nickel steel and low manganese steel. The results of the torsional strength at the various temperatures are shown in Fig. 37. The similarity between the two diagrams is quite striking and leads us to believe that if we had selected the short twisting test as a means of estimating the relative creep values of the different steels, we would have reached the same conclusions as those of Professor Norton, based on very long creeping tests. Both diagrams show that the nickel-chromium austenitic steels are distinctly in a class by themselves, all other steels being grouped together. They also indicate that as the temperature increases, the gap between the austenitic and non-austenitic steels becomes narrower and by interpolation one

might venture to predict that at 900 degrees Cent. (1650 degrees Fahr.) and at higher temperatures, that is when all steels are austenitic, regardless of their composition, they have substantially the same torsional strength and the same creep value, which, of course, would be very small. These views are offered here with full appreciation that further investigation may prove that they should be modified.

I feel that I owe you an apology for presenting the data of an obviously unfinished investigation. My excuse is the belief that it may, nevertheless, be of interest to some and that it may result in constructive criticism and in stimulating research along similar lines. Should we wait until an investigation is really completed we would forever remain silent.

Other features of interest observed in the conduct of these twisting tests might have been included in this lecture—but I have already trespassed on your patience and I shall close in thanking you for your courteous attention.

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PRODUCTION OF ELECTRIC STEEL FOR CASTINGS

BY GEORGE BATTY

Abstract

The methods applied in melting and finishing a heat of steel for castings must be considered as of primary importance because a casting cannot be expected to be better than the metal from which it is produced.

In this paper two sharply differentiated methods of procedure are discussed as being applicable to the production of both straight carbon and alloy steels for castings. It is affirmed that, in the ordinary conditions that pertain in a steel foundry, where scrap contaminated by adherent sand is used as part of the charge, a definite boil of the bath must be secured—with some inevitable elimination of carbon—in order that finely divided nonmetallics be cleansed from the bath.

Further, it is affirmed that to "kill off" the final traces of free and active oxide of iron in the bath with ferrosilicon is neither sound nor desirable practice.

IN this paper it is not the intention of the writer to propound any new theories in respect to steel making practice, but rather it is his desire to put before the members an outline of certain conditions that have to be met in the manufacture of steel castings and to stimulate a discussion on the subject of the making of steels for castings.

The advent and the increased application of alloy steels as castings has made them particularly satisfactory subjects for the attention of steel treaters, and it is largely due to carefully controlled heat treatment that the virtues of many types of steel castings are so developed as to make them a potent factor in commerce. Before treating the matter of the production of alloy steels in the electric furnace it will be well to review the methods that are used in producing plain carbon steels for castings.

There is not a uniformity of opinion as to the best method to be adopted in producing straight carbon steel for castings,—some makers hold that the best steel is produced by adjusting the components of

the charge to produce a certain carbon content without having to induce a heavy boil, or even an appreciable boil, to bring the carbon down to the desired figure. Other steel makers insist that the better method is to insure that every heat of plain carbon steel for castings be boiled, which involves the elimination of a considerable proportion of the carbon of the original charge. With the latter view the writer concurs.

The first method is practically identical with the practice adopted in the manufacture of high grade crucible steels and it may be that if the same care were exercised, or exercisable, in respect of the constituents of a charge to an electric furnace as was considered essential in the crucible charge, there might conceivably be some real justification for the opinion held by the protagonists of this method that it is competent to produce a superior steel. In general, however, it must be said that a steel foundry uses at least 30 per cent of its charge to the furnace in the form of scrap such as gates and risers, and such scrap is contaminated to a certain extent with sand adhering to the metal. In the opinion of the writer this feature alone is sufficient to demand the application of oxide of iron in some form to the bath in order that the steel may be cleansed.

Adherent grains of sand may be broken up into particles so fine that they are unable to rise through the liquid metal and become incorporated in the slag. It is not known to the writer that any foundry goes to the trouble of tumbling or otherwise cleaning the scrap before using it as part of the charge; hence there would appear to be the ever-present danger of finely divided silica in suspension in the metal of a charge which is melted, finished and poured without the introduction of such oxide of iron as will promote a boil. Included silica or silicates tend to reduce the ductility of the metal and certain failures of test bars may be identified with such nonmetallics. Others, of course, may be identified with silicates or oxides produced in the metal by certain reactions, but at the moment the consideration is limited to the crucible method of working a charge in the electric furnace.

In addition to the difficulty of ensuring a charge free from contamination by adherent silica, there is also a certain degree of trouble in securing other components of the charge free from scale or rust. Some steel makers refer to the method of melting at present under discussion as "melting with high residual manganese and silicon." Such procedure is possible if the whole of the components of the charge be mechanically cleaned before they are delivered to the fur-

nace, but is is very doubtful whether this is uniformly practiced by the people who advocate the crucible method of treatment in the electric furnace.

Any oxide of iron on the charge inevitably involves a reaction with the initial silicon and manganese of the charge and even though a perfectly clean charge be put into the furnace it is usually found that the material near the door or doors becomes heavily oxidized during the melting-down process through the induction of air at the door or ports. Attempts are made to seal the furnace and it is conceded that there is a possibility of protecting the charge from incidental oxidation during melting down, but such procedure is very rarely followed.

The condition, therefore, arises in the so-called "crucible" method, of having a small proportion of free oxide in the charge which will react selectively with the silicon and produce silica. The silica so produced will be in a very finely divided condition and therefore unlikely to levitate through the liquid steel to become part of the slag; here is seen a danger of the contamination of the charge with an undesirable nonmetallic constituent and, although the process may appear to be cheap in that it conserves silicon and manganese, it is open to serious criticism because of the liability of contaminating the charge with finely divided nonmetallics such as almost inevitably result in the ultimate product being depreciated in ductility. This point must be borne in mind when we come to consider the production of certain alloy steels.

The molecular silica formed by a reaction between silicon and a small amount of oxide of iron,—and the finely divided silica which adhered to the casting scrap of the charge,—is unlikely to be levitated to the slag because the tiny particles do not coalesce readily, but if we review the alternative method of producing steel in the electric furnace,—that is, by oreing down to a low carbon,—we insure a sufficient excess of oxide of iron to form fayalite ($2FeO.SiO_2$) which more readily coalesces into pieces large enough to rise to the steel-slag interface and become part of the slag.

It is admitted that in the process which involves oreing down to reduce the carbon to low limits, almost the whole of the initial silicon and manganese are converted to oxides. The reaction of iron oxide with silicon and manganese proceeds at lower temperatures than does the reaction of iron oxide with carbon; hence in the earlier stages of

the melt, the reaction of oxide of iron with certain constituents of the charge is selective in respect of silicon and manganese.

Insuring an excess of oxide of iron over the amount essential to convert the initial silicon and manganese to silica and manganous oxide is of primary importance in cleansing the steel of nonmetallics and it is considered desirable by the writer that the carbon content of a charge which is to finish at, say, 0.25 per cent, should be taken down, by oreing, to about 0.13 per cent or 0.14 per cent in order to insure, as far as is possible, the elimination of the products of the early reactions of oxide of iron with silicon and manganese.

It is probable that some steel makers who use the boiling down method negative, to a large extent, the value of the method through the manner in which they finish the heat,—after silicon, manganese and carbon are eliminated to the desired extent,—by killing off the remanent excess oxide of iron with small doses of ferrosilicon until the bath becomes quiet. The introduction of silicon to an over-oxidized bath will produce silica,—molecular silica,—which, in the absence of a sufficient amount of oxide of iron to form $2FeO.SiO_2$ will represent a serious contamination of the steel. It is, therefore, considered desirable to kill off the excess oxide of iron with some carburizing medium that contains very little silicon or manganese and for this purpose washed metal is recommended.

In an oxidized condition, a bath of steel is saturated with gas, but upon the introduction of a piece of cold scrap or pig iron there is an evolution of gas, presumably due to the physical effect of the material upon the saturated solution. The addition of washed metal, therefore, to an oxidized bath serves a double purpose in that it promotes a boil from purely physical action and continues the boil by supplying carbon with which the excess oxide of iron can react. This boil is evidence that the product of the reaction is gaseous and not solid, and it is the opinion of the writer that it is sound practice to prepare a heat of steel for the addition of the "finals" by so treating the bath with washed metal or other low silicon low manganese pig iron. The boil will help the levitation of any finely divided nonmetallics which were suspended in the bath and so will tend to clear the bath of such undesirable constituents. It must be remembered, however, that the material which is added as a carburizer should itself be low in nonmetallic inclusions,—as low as possible,—and that an excess should not be used, otherwise the carbon content of the bath

may be so raised that the addition of the finals,—ferrosilicon and ferromanganese,—will raise the carbon beyond the desired limits. After the addition of the carburizer a sample is taken and quenched for fracture examination, provided the melter is competent to judge with reasonable accuracy the carbon content of his material from a fractured sample. This examination should precede, in all cases, the addition of the finals in order that there may be a reasonable assurance that the material will conform with the chemical specification.

Some steel makers claim that they kill off a bath of steel, without the addition of ferrosilicon, by adding powdered coke or ground electrode to the slag. With this practice the writer does not agree because the carbon does not go to the metal and reduce the oxide of iron in the bath direct, but a reaction is set up between carbon and silica in the slag near the electrodes, whereby silicon is produced and precipitated into the steel. Under such procedure silicon is the killing agent and inevitably produces silica in the metal. Another deleterious feature of killing off a bath of metal with coal additions to the slag is that the steel may come to the condition which has been described as "over-reduced." It is not the writer's intention to deal with this aspect of steel making at the present time as it is probable that this subject will be treated in a separate paper at some future time when an investigation, now current, is carried further forward.

The ratio of slag to metal is quite an important matter in electric furnace practice and some operators are prone to have a slag which is far too heavy, that is, too great in weight and volume, in relation to the weight and volume of the metallic bath. It frequently occurs that a melter observes, shortly after the addition of ore, that the slag has become very thin and in some foundries it is fairly common practice for the melter to fear this condition and thicken up the slag with sand. Such procedure appears to be regular practice with some melters and is not justifiable if the whole of the facts are carefully considered. As a general rule the slag is thin because it contains an excessive amount of oxide of iron and, as this oxide has been put into the furnace for a specific purpose, it should be allowed to exercise its function uninterruptedly until the reaction has been carried so far that a suitably low carbon is attained. The dilution of the slag with sand decreases the concentration of iron oxide in the slag and so tends to slow down the speed of the reaction between iron oxide and carbon in the bath. Excessive use of iron ore is to be de-

precated, yet it must be admitted that many melters consistently use more ore than is necessary to complete the essential reactions because they have adopted the habit of building up a big slag by adding sand as soon as the slag appears to them to be too mobile. Instead of using a large ore addition, it is advantageous to use a smaller addition of ore and, as soon as this has been assimilated and a boil commences, to add limestone. Calcium is competent to replace iron to a certain extent in an acid slag; hence, by the use of limestone, we may limit the total amount of iron oxide in the slag and this is considered desirable. In open-hearth practice it is easier to secure a steady and regular carbon elimination when limestone is added to the slag than it is when no limestone is used. The same comment may justifiably be made in reference to acid electric practice. A limestone addition which will contribute a total of 8 per cent CaO to the slag is quite safe and such a slag is not so erosive to the banks as is a slag very rich in oxide of iron.

European practice seems to favor, in both acid open-hearth and acid electric, a high proportion of MnO in the slag and a low proportion of FeO. This is also considered desirable as applied in the manufacture of steel castings and there seems no good reason why a certain amount of slag should not be re-charged to the furnace in order to build up the MnO content of the slag where the charge is of such a composition that the normal MnO content of the slag would be under 14 per cent. A slag which is low in MnO involves the steel maker in rapid losses of manganese from metal to slag when the final additions of ferrosilicon and ferromanganese are made, and melting under a slag fairly high in MnO tends to conserve the manganese of the charge. Different types of arc furnaces appear to behave differently in respect of the elimination of manganese from the steel and the writer has no explanation to offer of this seeming peculiarity but would be interested to hear some discussion on this point.

The alloy steels to be discussed in this note are those which are deemed low alloy steels, containing a maximum amount of metallic elements, other than iron, of 4 to 5 per cent. Of these the most common is the nickel-chromium series. Remelting nickel-chromium scrap in the acid furnace inevitably involves a considerable loss of the chromium content of the scrap. With the oxidizing method advocated for plain carbon steels, this loss is inevitable, yet it is not suggested here that the "crucible" practice be adopted even though

chromium might be saved. The same criticism of the crucible practice applies with alloy steels as it does with plain carbon steels where foundry scrap is a considerable part of the charge and it is well to face the fact that certain elements are to be lost almost entirely from foundry scrap when remelted. Chromium, vanadium and manganese are readily oxidized and are lost from the metal to the slag. Titanium, if any, is residual in the steel and is readily oxidized. Nickel and copper do not suffer losses to any noticeable extent in remelting as scrap and molybdenum is very considerably recovered if the furnace is kept tight during the melting period. If, however, molybdenum-bearing scrap is allowed to become seriously oxidized before it is melted, there appears to be a definite loss of molybdenum through the volatilization of molybdenum oxide.

In the production of alloy steels, therefore, there is a greater proportion of loss on remelting scrap than there is on plain carbon steels and, in such scrap as contains chromium or vanadium in combination with other metallic elements, these two elements are almost a dead loss to the steel maker; hence, there may be a serious cost increment involved in such losses from remelted scrap. Because of the objections advanced against the crucible method of melting in the electric furnace, it is not deemed safe to advocate segregating alloy steel scrap and melting down a full charge of such scrap under conditions that would tend to conserve the special elements.

In making chromium-vanadium steels, the bath should be brought to good condition of suitable composition except for the special elements before the ferro alloys are added and these should be added in small pieces, if a small furnace is being operated, and worked into the bath with a bar. The slag condition immediately before the additions of these special alloys should be carefully adjusted, but the over-reduced condition previously referred to must be avoided so far as is possible. A time sufficient to insure the entire digestion of the ferrochromium by the bath must be allowed before the heat is tapped from the furnace, and if the slag condition is at all poor there may be extravagant losses of chromium to the slag. With vanadium the losses would be even more serious and some steel makers prefer to add ferrovanadium in the ladle rather than in the furnace, taking precautions to ensure that the slag is held back in the furnace until practically the whole of the metal has flowed into the ladle.

Molybdenum may be introduced at any time to the charge, either

as ferromolybdenum or as calcium molybdate. The latter is preferred and it is considered good practice to add the molybdenum when a good bath has been secured, but before the whole of the charge is completely melted. Nickel may be added at any time, but copper is usually added after the bath is completely melted.

There appears to be one danger attendant upon the use of copper but this is usually associated with a rather high phosphorus content and, in the writer's experience, the trouble has been most marked in basic electric steel. It would appear that copper enters into combination with phosphorus to form Cu_2P which is not dissociated in the subsequent melting operations and occasionally steels are produced which appear inherently brittle particularly in the heavy sections where the phosphide of copper has an opportunity to form comparatively heavy boundaries to the crystal grains. This is largely an assumption and may promote some discussion.

This note pertains primarily to the production of steel in the acid electric furnace and it is not intended here to inject anything particularly relative to basic practice but there seems no good reason why such a subject should not be touched on in the discussion.

Aluminum. In every open discussion in which the writer has taken part in this country where steel castings were the subject under consideration, reference has been made to what is called the "low ductility problem" which, in particular, is identified with green sand castings. The Steel Castings Development Bureau, through their associate at the Bureau of Standards, have in hand an investigation of this problem and, while such investigation is not yet completed, the writer feels competent to express the opinion that the low ductilities are not primarily a metal problem insofar as melting operations are concerned. This is said because it has repeatedly been proved that metal which is competent to produce perfectly sound castings when poured in dry or skin-dried molds will not produce equally sound castings from straight green sand molds. The metal is identical as it enters each mold; hence the variant must be the mold. The only variant as between the skin-dried green sand mold and the straight green sand mold is the moisture content of the sand at the face of the mold; hence we arrive at the conclusion that moisture is the primary cause of the pin hole trouble in green sand castings.

Founders rapidly came to the conclusion that aluminum cured the pin hole trouble. In such an assumption they were correct, but

they were involved, in curing one trouble, in another trouble; namely, low ductilities.

Quite a striking demonstration of the origin of the pin hole trouble is provided in the test mold which is standard in the foundries of the Steel Castings Development Bureau. This is illustrated in Fig. 1 which shows a section cut through a mold and test block. This sketch is truly indicative of the conditions that pertain and a sketch is used instead of a photograph in order that the mold design may be shown. One shoulder is produced in a dry sand core, the remainder of the mold being green sand. Metal poured into this

A.   Dry sand core
B.   Green sand
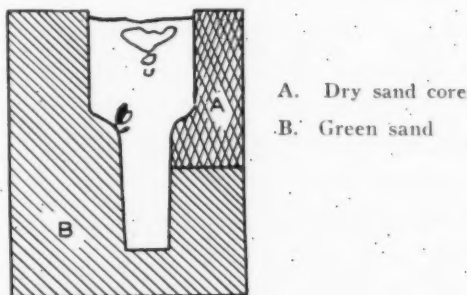
Fig.   1 — Condition
Indicative of Unsuit-
able Green Sand.

mold provides evidence of the value of the green sand as a molding medium and as such is a much more useful indication of sand condition than the more usual control methods of permeability tests and moisture determinations. It is, however, used in combination with, and to amplify, such tests. Fig. 1 shows a condition indicative of unsuitable green sand and indicates the fact that the metal is constitutionally good because there are no blows or pin holes from the dry sand shoulder of the mold, whereas from the green sand shoulder of the mold, where this part of the mold is disproportionately heated, there is an evolution of gas—largely hydrogen and oxygen from steam which is split up by the heat of the metal—that penetrates the initially solidified envelope and forms pin holes.

The pin hole problem and the problem of low ductilities are, in the opinion of the writer, inevitably related to the moisture content of the sand and it is from this angle that the problem is being attacked. At the same time, numerous experiments are being conducted with deoxidizers other than aluminum with a view to finding a medium which, upon reaction with immigrant oxygen produces an oxide

that is less deleterious than is alumina. It is not safe to attribute the whole of the reduced ductility to locally formed alumina, yet there can be no doubt that such alumina is a very potent factor, probably the main culprit, in producing the less ductile condition.

The injection of this matter of the use of aluminum is justified in the opinion of the writer in order that users of steel castings may be made aware that steel founders are not accepting present conditions as representing the bounds of possibility, but are earnestly investigating the problem from all significant angles with a view to producing castings that are entirely dependable.

Another point to be stressed is, that because certain castings of intricate design which have to be produced in green sand molds have shown ductilities of the order of only 18 per cent elongation instead of 24 per cent there is no warrant for any assumption that this condition is representative of steel castings as a whole in that particular range of composition indicated by a particular specification. The condition has arisen primarily because the design of the casting is such that nothing but a green sand mold is a suitable medium to produce it free from cracks or checks. The low ductilities occasionally noted are associated with the fact that free aluminum must be added to the metal in order to react with immigrant gases from the mold and eliminate such blow holes, referred to as pin holes, as would be formed at re-entrant angles of the mold. The steel itself is inherently of physical properties superior to those revealed in a test bar produced under identical mold conditions with the casting or test bars cut from the casting, and quite a large proportion of the castings—the heavier type of castings such as have to withstand arduous service—are not subject to the same mold conditions and are, therefore, comparatively immune from the deficiencies of light castings of intricate design.

Within the past year quite a lot of work has been done on this subject and distinct advances have been made in defining mold conditions and reducing the incidence of the pin hole trouble, thereby reducing the incidence of the low ductility trouble. Resultantly castings of higher quality are being produced purely as a result of protecting the steel, after it enters the mold, from depreciation by immigrant gases or by depreciation through the local formation of oxides of such metals as were introduced to react with the immigrant gases. Steel-making procedure is constantly under close scrutiny and

it is felt that the advent of scientific control in all stages of manufacture is tending rapidly toward the production of superior steel castings—both plain carbon steels and alloy steels—that are a suitable medium for the exercise of the art of the steel treater.

DISCUSSION

Written Discussion: By G. F. Comstock, Niagara Falls, New York.

Mr. Batty's remarks on porosity and the effects of aluminum in steel castings were especially interesting to the writer, who would criticize the statement that "alumina is a very potent factor, probably the main culprit, in producing the less ductile condition" in steel castings treated with aluminum. In the metallographic laboratory of the Titanium Alloy Mfg. Co. a large number of specimens of cast steel from a great many foundries have been examined by the writer, and studied with special reference to the correlation between the inclusions in the steel and the tensile test results. Steels treated with aluminum almost invariably showed more or less alumina in the polished section, but hardly any inclusions of silicate slag. Steels not treated with alumina, on the other hand, showed silicate globules but no typical alumina. The aluminum-treated steels, as a class, were invariably inferior in ductility to those not treated with aluminum, as stated by Mr. Batty. Except in those instances where the alumina was badly segregated, there did not seem to be any good reason why an alumina inclusion should be any more effective in reducing the ductility than a slag or silicate inclusion, and there were often plenty of the latter kind in steels of excellent ductility.

After several years' study of this question, the conclusion was arrived at that the difference in ductility between these two types of cast steel was due to the different arrangement of the sulphide inclusions. In the aluminum-treated steel the sulphides are practically always very fine, and arranged more or less distinctly in streaks forming a sort of fragmentary network. Steels cast without aluminum normally have their sulphide inclusions coarser, globular, and more scattered, like the silicate globules. In the tensile test the streaks of fine sulphides have a tendency to cause cracks to open up in the specimen after the yield point is passed, and these cracks grow under the increasing load so that the specimen breaks with less elongation and reduction of area than if the cracks had not opened. Inclusions scattered and not arranged in streaks do not have such a strong tendency to cause cracks to open up in a stretching test bar. We have found that low ductility could be correlated with the streaky sulphide arrangement more closely than merely with the presence of segregated alumina inclusions, although as stated both these characteristics often occurred together.

The reason for the streaky sulphide arrangement being found chiefly in steels treated with aluminum is not entirely clear to the writer. It is a fact however that we find it nearly always in cast steel so treated, and
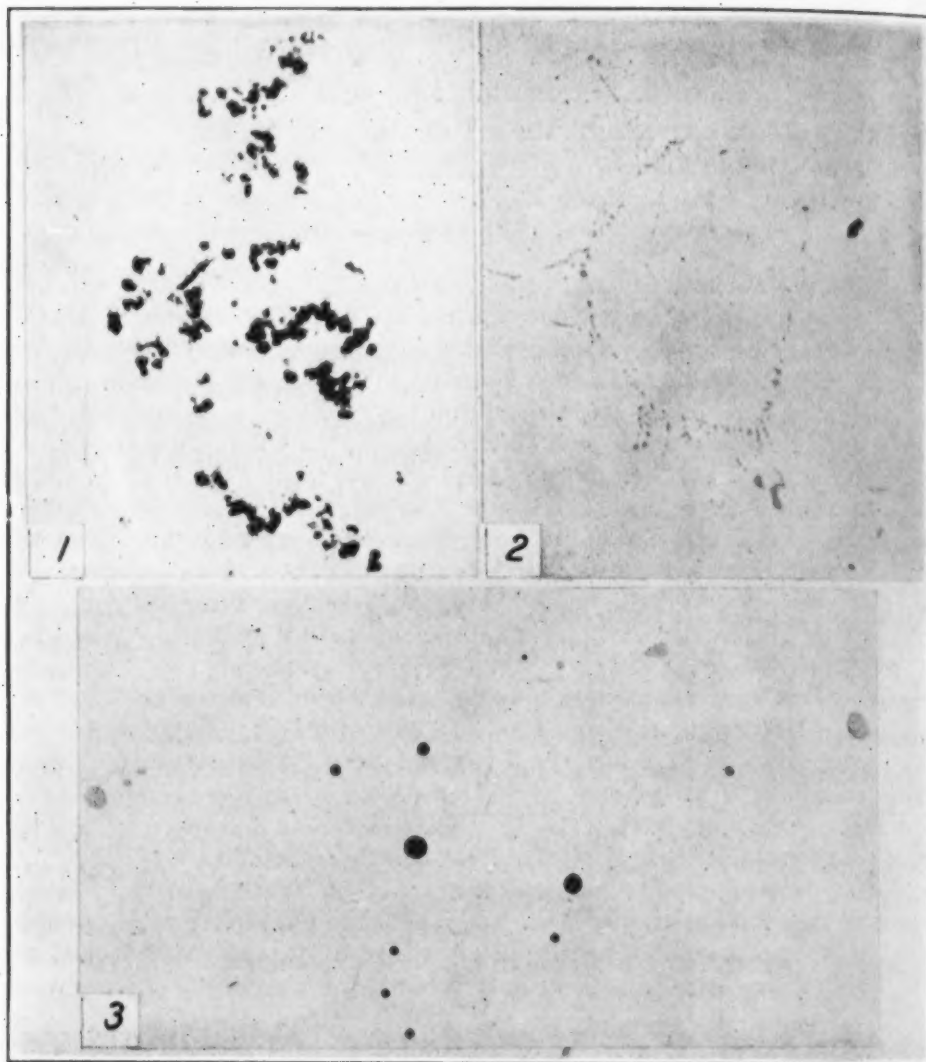
Fig. 1—A Group of Typical Segregated Alumina Inclusions in a Cast Steel Treated with Aluminum.
Fig. 2—A Few Dark Particles of Alumina, and Fine Light-Gray Sulphides Arranged in the Network Pattern Typical of Aluminum-Treated Cast Steel.
Fig. 3—Dark Silicate or Slag Globules, and Light-Gray Sulphides in Similar Scattered Globules, as Found Normally in Cast Steel Without Aluminum. All Photomicrographs × 200, Unetched.

only seldom in other cast steels. The subject was discussed in more detail in *Iron Age*, Dec. 4, 1924, p. 1477.

The writer would like to make a plea in this connection for a more careful study of the inclusions in steel, especially by such Bureaus as the author directs. When studying the effect of aluminum, one should not assume that all inclusions in the steel studied are alumina; nor when studying the effect of sulphur, that they are all sulphides. Too often the nonmetallic particles in steel are considered as merely "inclusions." It

is a simple matter, with reasonable care, to classify the inclusions as sulphides, silicates, alumina; etc., and there would seem to be little excuse nowadays for the failure to make this distinction.

Three photomicrographs are presented to illustrate some of the types of inclusions that have been described.

Written Discussion: By R. A. Bull, director, Electric Steel Founders' Research Group, Chicago.

Most of the experienced open-hearth and electric melters will endorse Mr. Batty's opinion of the advisability of having a boil in the bath, brought about by the introduction of iron oxide in some form. Indeed, there are numerous open-hearth and electric melters who are unwisely addicted to the habit of providing for an excessive boil, in the present writer's opinion. His observations lead him to believe that excellent and possibly the best results are obtained from a boil that has what may be termed a gentle behavior. One reason for this lies in the fact that the regular practice of providing for a very vigorous boil as compared with a gentle boil automatically leads the average melter occasionally if not frequently to introduce some iron ore at too late a period in the working of the heat.

The present writer has just indicated that he has what some may call prejudice against belated iron ore additions. He confesses to his serious objection to the introduction of iron oxide except during the comparatively early stages of the working of the heat.

It would be extremely difficult in the case of any particular furnace and customary heat charge to determine exactly how long it takes on the average, for the oxides to be worked out of the heat so as to leave the metal and the slag in suitable condition preparatory to tapping. We may perhaps generalize to some good purpose in the case of the ordinary acid electric furnace utilized for the manufacture of steel for castings. Most such furnaces are of three tons rated capacity, and probably the majority of them customarily receive cold charges around 7000 pounds, or at least a charge somewhat in excess of the normal capacity as designated by the furnace designer.

The present writer admits that he theorizes, perhaps to a hazardous extent, in expressing his personal opinion that there should be no introduction of iron ore into an acid electric steel heat of the size just referred to, later than 45 minutes before tapping time. This opinion is advanced without the opportunity of definitely determining what period of time ordinarily is required to eliminate from the bath of metal, the removable oxides introduced with the ore. Observation of conditions in the steel poured into casting molds presents the only justification for the recommendation. If this recommendation should fail of general acceptance in any present or future discussion of the subject, there may at least accrue some advantage from it, if it prompts a broader exchange of ideas and a more serious consideration of such steel foundry conditions as are observable and have practical significance. This being so, the present writer feels free to express the opinion he now holds, reserving for him-

self the right subsequently to modify his views in this as in other details according to evidence that may be accumulated.

Mr. Batty's references to the influence of copper are interesting. He advances the idea that cold shortness that appears to him to be associated with a copper content in the steel may be dependent to considerable extent on the combined influence of the element mentioned, and phosphorus. When the percentage of phosphorus is fairly high and when there is an appreciable proportion of copper in the metal, the present writer believes the observable effect on the product would essentially be red shortness, due to both copper and phosphorus, and cold shortness, due to phosphorus. If copper is a contributory cause for cold shortness, the present writer has not observed definite evidence of the fact. He believes, however, that phosphorus is a contributory cause for red shortness—not by any means to the same degree that sulphur demonstrates its effects, but to some extent, as indicated by conditions observed in the manufacture of certain steel castings that by reason of their designs were susceptible to shrinkage cracks, and which therefore called for the maximum precautionary measures in the manufacture of metal and mold.

It is generally conceded that copper in large quantities has a red shortening effect, the bad results of which are by the nature of the case felt by the foundryman, and not often by his customer. At what minimum percentage this influence begins to be exerted is probably not known. The increasing difficulty of obtaining purchase scrap for electric steel manufacture, free from alloying elements outside the scope of what is termed common steel, points to the advisability of ascertaining what effects are produced in casting steel, not only by copper but by other elements which frequently exist in the scrap charge, and which are present without being desired by the foundryman. The degree to which the use of alloy steels is being extended makes quite probable the utter impossibility, five years hence, of obtaining such material in a physical form most suitable for electric steel melting, and at the same time having a chemical composition justifying classification as common or carbon steel.

The present writer subscribes unreservedly to the author's opinion that porosity in steel castings is caused generally by mold rather than metal conditions, although there is no disputing the fact that pin holes may be produced by an undesirable condition of the metal. No one who is even partially informed on the subject would take issue with Mr. Batty's references to the injurious effects on ductility of aluminum. In saying this the present writer does not take issue with those who have pointed out from time to time that an appreciable amount of aluminum added to a well prepared steel heat appears in occasional instances to cause no bad effects on the physical properties.

It is the present writer's belief that some persons have been deceived as to the influence of the aluminum by failing to consider the time factor. When this deoxidizer is added to the metal at the furnace spout during the tapping operation, and the metal subsequently is permitted to acquire and remain in a quiescent state, ultimately being poured into

molds by withdrawal of the steel from the bottom of the original ladle used, there appears to be the minimum opportunity for harmful oxides to find their way into the castings. Time permits the lighter inclusions to rise toward the surface of the metal, when there is no mechanical disturbance to interfere. For these reasons a moderate aluminum addition made to a ladle of steel at the furnace spout may very often fail to result in apparent injury to the steel, if a sufficient period is allowed for the inclusions to levitate, and if pouring is done through a nozzle at the bottom of the ladle. However, there is a disturbing circumstance in the demonstrated fact that the aluminum-treated steel is effective in eliminating the mold gases, practically in proportion to the lateness of introducing this element. Very hot steel exposed for a short interval to the atmosphere before it enters a green sand mold is unquestionably in better condition to eliminate the mold gases when the aluminum is added just before the mold is poured, than when the aluminum is added at the furnace spout.

The research for a satisfactory substitute for aluminum is thoroughly warranted. It has been carried on by many steel foundrymen, but with a degree of success that to large extent is indirect rather than direct. Most of those who have earnestly endeavored to find such a substitute have probably reached the conclusion that the best results of efforts to prevent pin holes in green sand steel castings and at the same time to employ judiciously a very small proportion of aluminum consist in measures to prevent the condition which aluminum corrects, namely, an excessively gaseous condition in the mold when it is being poured. This is being accomplished to considerable extent by carefully determined and positively maintained control procedures, related not only to the all-important matter of moisture in the molds, but to the phenomenon of volatility as involving certain relatively dry materials in the mold; to venting practices by means of channels mechanically created to permit the exit of gases; to the permeability of the molding sand itself; to the suitable condition of ladles and all other equipment required for heat pouring; and in brief, to every one of a very large number of details of foundry shop practice which may be contributory if not prime causes for the formation or retention of gas.

Written Discussion: By Frank Hodson, president, Empire Steel Castings, Inc., Reading, Pa.

The actual method to be adopted to achieve the best day to day results is a matter of great importance and personally I am of the opinion that a balance can be struck between the advocates of a purely crucible type melt and the heavy boiling down of carbon advocated by Mr. Batty. This undoubtedly involves a closer personal attention to the melt and a better type of furnace operator than is usually met with in steel foundry practice but surely it is false economy to put the most important part of an electric furnace plant operation in the hands of anybody but a skilled, well trained practical metallurgist. Mr. Batty, in the preamble of his paper, very rightly stresses the primary importance of melting

and finishing a heat—"because a casting cannot be expected to be better than the metal (meaning of course the molten metal, not the cold charge) from which it is produced." Subsequent molding, pouring and the treatment the metal receives after it leaves the furnace spout will not improve what the melter produces and in most, if not all cases, it will not be as good as the metal in the furnace. We take great care throughout the melting and finishing to avoid over oxidizing the steel, or from allowing air to impinge upon the bare metal, and yet the molten metal in every steel making process has, during the pouring, to pass at a very high temperature through air, and also to come in contact with refractories, sand, and so forth, that probably change its whole physical characteristics. It is certain that two steels made from the same mix or of the same composition, one of which is poured or allowed to cool in vacuum and the other poured in the usual manner will show entirely different properties—particularly as regards ductility and toughness. Nitrogen from the air may be the answer but it is merely given as a thought for future investigation and research, and an illustration that the presence of nonmetallic inclusions of silica, alumina and so forth—whilst admittedly impairing ductility—may not be more important in a correctly melted electric steel as other things now ignored.

Mr. Batty points out certain castings bought under Government specifications are so intricate that only green sand molds can be used, and to overcome the inherent tendency of such castings to pin holes from re-entrant angles of the mold, aluminum must be used in excess. Would it not be a good thing to suggest to, say the Navy Department inspectors—that in such cases an elongation of 18 per cent in a test bar might denote a sounder stronger actual casting, than 24 per cent elongation in a test bar treated with less aluminum. Provided the inspection departments are satisfied a casting should be made in a green sand mold, then the present insistence of high elongation with no reference to possible higher elastic limit and breaking strain are merely an unfair handicap on the foundries making good steels.

Our own experience on remelting of alloy steels—particularly the chromium steels, in an acid furnace—would show that with proper knowledge and care the losses can be kept down between 2 and 3 per cent of the chromium introduced in the scrap. However, the personal equation may enter considerably into this.

Written Discussion: By H. D. Phillips, Empire Steel Castings, Inc., Reading, Pa.

The author's statement concerning the difference of opinion as to the best method to be adopted in producing straight carbon electric steel for castings is generally accepted. At the same time, it is my opinion that the best method to be used in their manufacture is the so-called "crucible" practice of melting and refining. The method herein described is perhaps a modification of the practice avoided by Mr. Batty, being, however, a distinct improvement over the same.

The care exercised in selecting the scrap to be used in the crucible

manufacture of high grade tool steels need not be exercised in this prac-- tice. It might be stated here that care should be exercised in the selection of raw materials entering into the charge used in any of the various melting practices, this care being of absolute necessity if a uniform melting practice of any manner is to be attained. The only care, therefore, exercised in this practice is that used in all practices in which uniform metal is desired. It is not necessary to obtain a charge free of rust or scale in order to have the carbon content in the desired range when the heat is tapped, without inducing a boil, or recarburizing during the heat. Melting with a high residual manganese and silicon need not be practiced as this is only possible, as Mr. Batty states, when the metal charged has been mechanically cleaned before using, and even if this practice were followed, the amount of iron oxide formed during melting would react with the manganese and silicon in the charge, affecting a very noticeable reduction of these elements. As the author states, the practice of sealing the doors during the melt-down is very rarely followed. It is, however, practiced in this case in making charges that lie in the furnace overnight. The first heat is always charged the day preceding and it is possible, by using care, to prevent the formation of considerable quantities of iron oxide in the charge lying in a hot furnace overnight. The main reason for not sealing the furnace during the melt-down of each heat is the extreme rapidity with which melting operations are carried on in a steel foundry and, also, as a further reason, the limited furnace crew employed, notwithstanding the slight advantage gained in insuring a higher manganese and silicon content in the bath formed. It is agreed that the finely divided nonmetallics formed during melting seriously impair the ductility of the steel unless removed, although it is not in my opinion necessary to form fayalite ($2FeO.SiO_2$) in order to insure the removal of these sonims or nonmetallic inclusions in the steel. The formation of fayalite is only possible with an excess amount of iron oxide present in the bath and to attempt to kill an over-oxidized bath with additions of ferrosilicon will result in an excess amount of these deleterious inclusions being present in the finished steel. It is true that an over-oxidized bath should be brought back to a normal condition by a deoxidizer whose product of reaction will be a gaseous instead of a solid substance.

Unless the bath is allowed to react for a considerable time with the additions of washed metal there will still be certain amounts of iron oxide present, and it may be safely stated that it is only theoretically possible to remove all traces of iron oxide from the bath by the additions of carbonaceous materials, and any iron oxide remaining promptly reacting to form inclusions of silica upon the addition of ferrosilicon being made to the bath. As can be seen the small amount of iron oxide present reacting with the silicon added again charges the metal with finely divided particles of silica. It would really seem to be a waste of time to bother ridding the bath of these inclusions at so great an expense as in the practice advocated by Mr. Batty, merely to have it charged with more inclusions of like nature upon adding the carburizing materials. Also the

material used, washed metal or low silicon low manganese pig iron, in bringing the carbon up to the desired content can reasonably be expected to contain some nonmetallic inclusions.

In the writer's experience, the crucible method of melting produced a high grade of steel, showing good ductility for a plain carbon cast steel. The metal under the microscope was remarkably cleansed of nonmetallic inclusions, and in most cases was considerably cleaner than metal made by the method concurring with Mr. Batty's opinion. The care used in selecting the scrap has been explained before, as uniform a practice as possible being attained, each heat being as near as possible a duplicate of the preceding one. No attempt was made to melt with a high residual manganese and silicon, though care was exercised so that too severe oxidation did not occur. The slag was reduced by means of coke or charcoal, limestone additions and also additions of bauxite or loam being regularly made in order to keep the slag in a fluid condition. No attempt was made to reduce the silicon from the slag, the carburizers being added when the slag, hence the metal, was as free as possible of any iron oxide. As it is the nature of an electric furnace of the arc type, to keep the bath in a constant state of motion, the levitation of silica inclusions is aided considerably without the necessity of violently boiling the bath. A slow working action is maintained in the bath by the additions of limestone which also aids in the purification of the steel. No appreciable loss of carbon is attained. In making the charge a few shovelfuls of limestone are placed over the charge when about one-third completed. This limestone aids considerably in the fluxing of silica inclusions formed during melting. The calcium oxide (CaO) content of the slag was high, there being also a high proportion of manganese oxide (MnO) present. The additions of ferromanganese and ferrosilicon are made simultaneously, primarily to aid in the coalescing of the nonmetallic inclusions formed. With the bath well deoxidized the losses of silicon and manganese are slight, therefore, the liability of contamination by the products of the oxidation of these two elements, is slight. In the practice recommended by Mr. Batty, the slag and the metal being in a fairly oxidized state, the loss of these elements can reasonably be expected to be higher, causing the recontamination of the metal, as has been explained before. As a final discussion of this practice used in making straight carbon steel for castings, it can be stated that a uniformly high quality product was produced.

In discussing alloy steels, it has been possible to duplicate heats of chromium steels in which a maximum loss of 2 per cent of total chromium was attained, merely by following the same procedure as used in the melting of ordinary carbon steels. This loss of chromium takes into account all chromium present including that charged in the return scrap which made up approximately 35 per cent of the total charge. It is an unnecessary waste to deliberately set out to lose contained chromium or other alloys in the heat and attain the desired composition by the addition of these same alloys near the finish of the heat, involving a further

loss if the slag is at all poor. To make a high carbon or alloy steel by the method advocated by Mr. Batty is an expensive proposition.

Coming to the question of aluminum, which is indeed a question to steel foundrymen, it is true as Mr. Batty states that the low ductilities present in small intricately designed steel castings made in green sand molds are due to the additions of free aluminum made to the metal reacting with the immigrant gases generated within the mold, forming inclusions of finely divided alumina. These inclusions of alumina seem to exert an interference to the slippage or sliding of the grains of the metal during times of stress, causing the low ductility. The statement concerning the low ductility caused by alumina is only applicable when metal from the same heat cast in a dry sand mold without having been treated with aluminum, shows itself to be a well made ductile steel. If this steel is not of a ductile nature the trouble probably lies in the melting or perhaps molding practice, as the metal in pouring into the mold sometimes picks up and holds small particles of sand, which is essentially a silica inclusion. It seems plain that low ductility in the metal can also be caused by poor molding and pouring practice. Coming to Mr. Batty's statement that "low ductilities are not primarily a metal problem insofar as melting operations are concerned," it would seem that any work done to cleanse the steel from silica inclusions formed during melting is not of the importance stressed earlier in the paper, as it seems from the author's latter statement that the primary factor in producing tough steels lies elsewhere.

J. A. DeBONDY: I would like to know if iron ore is used only if the carbon in the steel is too high in the melting; also, do you recommend the use of iron ore for a boil in all heats, and if iron ore is not to be used on every heat, how do you know whether iron ore is needed 45 minutes before tapping the heat.

Author's Closure

The written discussion by H. D. Phillips is a claim for the superior qualities of steel from the acid electric furnace, melted under conditions of control which involve the oxidation of most of the manganese and silicon of the charge. It cannot, therefore, be considered as identical with the "crucible" method of melting, and in my experience it does not produce metal of superior physical properties to that melted under oxidizing conditions involving the elimination of some carbon by a boil.

I cannot agree with Mr. Phillips that the elimination of nonmetallics produced by the oxidation of silicon and manganese is more complete without a moderate boil than it is when a boil is definitely induced and the carbon content of the bath reduced to the figure requisite for the production of steel of correct composition.

When an appreciable portion of the charge consists of foundry scrap and the remainder is made up of material of an average 0.25 per cent carbon content, the average carbon of the charge will be appreciably higher than the figure which must be attained in the melted bath before ferromanganese is added to replace the manganese which has been oxidized in the melting process. There is almost inevitably involved some carbon elimination because the

carbon increment from ordinary commercial ferromanganese is of the order of 0.06 per cent carbon to the bath. The charge must, therefore, comprise oxygen in some form of combination in a quantity sufficient to set up the selective reactions with silicon and manganese and then react with carbon, which latter reaction inevitably produces a boil. The practice advocated by Mr. Phillips, therefore, is that excessively high average carbon content charges should be avoided and in this I agree, but I certainly do not believe that a bath which will melt out at 0.20 per cent carbon necessarily produces better steel than a bath which melts out at, say, 0.30 per cent carbon.

From the standpoint of economy it is better to arrange the charge, by the purchase of proper materials, to melt out only a few points of carbon higher than the minimum figure to which the bath must be taken in order to produce the correct carbon content in the finished material when taking cognizance of the carbon increment from the ferromanganese.

Mr. Phillips operates in one of the foundries which comes directly under my scrutiny and I have no circumstantial evidence which would support his claim that metal produced in the way he indicates is in any way better than that produced under definitely controlled oxidizing conditions by other members of the organization.

In reply to the comment late in Mr. Phillips' contribution in reference to low ductilities, I would say that low ductilities are not primarily a metal problem insofar as melting operations are concerned because we have absolute proof in his own foundry, as in the foundries of the other members, that the care exercised in melting produces a steel which is satisfactorily ductile unless it is contaminated with alumina or with the products of a reaction between aluminum and immigrant gases from a green sand mold.

The author of a paper most surely has the right to object to any paraphrase of his written matter or to any misconception by any reader of the paper and it is here suggested that a short excerpt from a paper or even from a sentence must be considered in relation to the context if a correct appreciation of the words is to be attained.

It is my conviction that the problem of low ductilities such as are occasionally revealed in tests on green sand castings is one that primarily concerns mold condition—the metal is established as being of satisfactory composition and constitution, but is vitiated by the products of reactions between immigrant mold gases and components of the metal. From the "angle" of the mold this problem is being steadily worked upon and, whilst it is admitted that the temperature of the metal is also a significant factor, the mold composition and constitution remains the chief problem.

In reply to Mr. Hodson's written comment, I would say that I do not advocate a heavy boiling down of carbon, nor have I ever advocated such a high carbon charge as would necessitate a profuse boil in order to reduce the carbon to the desired limits within a short time. It is generally conceded that the influence of a steady boil helps in the elimination of nonmetallics suspended in the bath and if a bath is not allowed to boil, I very much doubt whether the resultant metal will be even reasonably freed from microscopic and submicroscopic particles of silica and silicates.

The method of finishing a heat of steel by using carbon in the form

of washed metal to remove the active oxide of iron still remanent in the bath when the carbon has been brought down to a sufficiently low figure is primarily designed to prevent the formation of such silica and silicates as normally result from the general practice of quieting a bath with silicon. It is not claimed that this carbon addition will remove every vestige of iron oxide from the bath, but in regular practice it is found that it so reduces the iron oxide content of the bath that perfectly sound steel free from blow holes may be produced, and that this steel is of satisfactory physical properties in every way. It should be obvious that any means taken to avoid the late formation of minute particles of nonmetallic matter which may remain suspended in the steel is a move in the right direction.

With reference to Mr. DeBondy's question, I would say that, as a general rule, the melter has an approximate idea of what his charge is likely to melt out at with respect to carbon. He has some ideas to the average carbon content of the charge and he will know whether or not he is likely to require iron ore to bring that carbon down to the desired point with an allowance made for the carbon—from ferromanganese, which is added to the bath in making plain carbon steel. Hence he will, as soon as he gets a good bath of metal, that is when the center of the charge is melted, add the requisite calculated amount of ore to bring the carbon down to the desired figure in the fully melted charge. You must realize that any reaction between oxide of iron as ore or rust on the scrap is selective with relation to silicon and manganese before it can attack carbon. Silicon and manganese are oxidized to a considerable extent before the bath is of a sufficiently high temperature to promote the reaction between oxide of iron and carbon. I think that should answer the first question—that iron ore is added to the bath to reduce the carbon to the desired amount and that silicon and manganese are oxidized previously because of selective reaction at a low temperature.

In plain carbon castings, as a general rule, a minimum manganese content of 0.65 per cent is specified. This demands an addition of ferromanganese which involves a considerable carbon increment to the bath; hence the carbon of the bath must be brought down to a figure appreciably lower than that specified, or necessary, in the finished charge. The oxidation of silicon and manganese in the early part of the fluid life of the bath produces nonmetallics and I think Dr. Herty could add very usefully to the discussion with his assertion that nonmetallics are eliminated from the bath to a more considerable extent if the carbon is taken fairly low. It has been my practice in making mild carbon steel for castings to run the carbon in the bath down to 0.12 or 0.14 per cent before making the final additions and to kill off the bath before making the additions of silicon and manganese with washed metal or, in England, with a good quality of Swedish pig iron low in silicon and manganese.

I consider it essential that a bath should boil when a fair amount of foundry scrap is a component of the charge that is ultimately to produce steel castings. The small inclusions which appear to contribute considerably to low ductility are particles a good deal finer than ordinary silica flour and adherent grains of sand break up, under the influence of the arc, into very

fine particles—particles so fine that they cannot levitate readily to the surface and become incorporated in the slag. These particles of silica need something to make them coalesce and, in normal melting, this is partially provided by the manganese of the initial charge but we have a greater assurance of the levitation of the nonmetallics when we have the mild boil condition because we are then certain that oxide of iron is present in sufficient quantity to form a silicate, which silicate levitates much more readily than does pure silica, because the silicate has a much greater tendency to coalesce.

It is essential that a bath be seen to boil—not violently, however. If, when the carbon content of the charge has been reduced to a sufficiently low figure, you have excess oxygen in the bath, this excess oxygen or oxide must be removed. If silicon is used to kill off this excess oxygen, silica will be formed and become a nonmetallic inclusion in the steel.

As a comment on Major Bull's contribution—I quite agree with him that it is undesirable to make a late addition of either ore or limestone to the charge but would make exception in certain circumstances. The circumstances to which I refer are only touched upon very briefly in the paper—the condition which I have called "over-reduction." I believe, and my opinion is substantiated by the physical test results we have had, that it is essential a bath be brought back to good condition from the over-reduced condition by a small addition of iron ore if we are to produce metal in a satisfactory state for pouring light and intricate castings. We must get away from this sluggish condition where the metal films over when exposed to the atmosphere in the handshank. Quite a small amount of iron ore, six or seven pounds in a 5000-pound heat, is sufficient to revert the metal from the over-reduced state to the fluidity which should be normal for the temperature. Any increase in nonmetallics is very slight and the physical qualities of the steel are not impaired to any extent so far as we have been able to determine.

Like other heats of steel, the physical properties as revealed by the test bars are some above the average, and some below the average. Over-reduced steel appears to have a greater linear contraction than does the normally reduced steel. That would indicate, I think, a higher freezing point and would predicate greater wastage from cracked castings. The lower fluidity would involve waster castings from incomplete running and it is therefore essential in making light steel castings that we have a high degree of fluidity so that the metal may run the casting cleanly with a minimum of wastage.

On the question of pin holes I maintain that these are, in the huge majority of cases, due to the immigrant gases from the molds.

With reference to Mr. Comstock's contribution, I would express my great appreciation and the hope that I may go further into this matter with him. The relation of aluminum to the sulphide inclusions may lead us to some quite important conclusions and into different avenues of attack upon the low ductility problem. At the present moment I can only offer to him my thanks for his contribution to the discussion as I am unable to express a substantial opinion upon the point he raises.

Recommended Practice Committee Releases*

STRUCTURE OF TIN-ANTIMONY-COPPER ALLOYS USED AS BEARING METALS

By O. W. Ellis[1]

BY referring to the data sheet dealing with the constitution of the tin-antimony-copper alloys it will be seen that the alloys of this system, which are of industrial importance, may be divided into four groups as follows:

1. The first group of alloys, as cast, are characterized by a hypo-eutectic structure, comprising alpha solid solution and a eutectic, or pseudo-eutectic, which is composed of the alpha solid solution and the intermetallic compound, $CuSn$, of the copper-tin system.

2. The alloys in the second group, as cast, contain three micro-constituents: the intermetallic compound, $CuSn$, the alpha solid solution, and the pseudo-eutectic.

3. The third group of alloys, as cast, comprise three micro-constituents: the intermetallic compound, $SnSb$; the alpha solid solution; and pseudo-eutectic. The proportion of pseudo-eutectic in these alloys is very small.

4. The fourth group of alloys, as cast, comprise four micro-constituents: intermetallic compound, $CuSn$; intermetallic compound, $SnSb$; alpha solid solution; and pseudo-eutectic, of which the proportion is very small.

These four groups lie within the area outlined in Fig. 1. Group 1 occupies the area ABDC; group 2, the area CDFE; group 3, the area BGHD; and group 4, the area DHJF.

The microstructure of typical members of each of these groups is shown in Figs. 2 to 7. In all cases the alloys were poured at a temperature of 400 degrees Cent. into a vertical mold approximately 1¼ inches in diameter and 7½ inches deep, which was heated to a temperature of 100 degrees Cent. prior to casting. The samples for micro-examination were cut from the castings at a point about ½ inch from the bottom. The section photographed was near the center of the casting.

Fig. 2 is the structure of an alloy containing approximately 95.5 per cent tin, 4 per cent antimony, and 0.5 per cent copper. It may be taken as representative of the structure of the alloys of group 1, which lie within the area ABDC of Fig. 1. The lighter areas are sections of primary

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alpha solid solution of the tin-antimony system; the darker areas are sections of pseudo-eutectic. The small white crystallites noticeable within the darker areas are of the intermetallic compound, CuSn, which is itself a constituent of the pseudo-eutectic.

Fig. 3 is the structure of an alloy containing approximately 94 per cent tin, 2 per cent antimony, and 4 per cent copper. This may be considered as a representative structure of the alloys of group 2, which lie within the area CDFE of Fig. 1 (see also Fig. 4). In this photomicrograph the
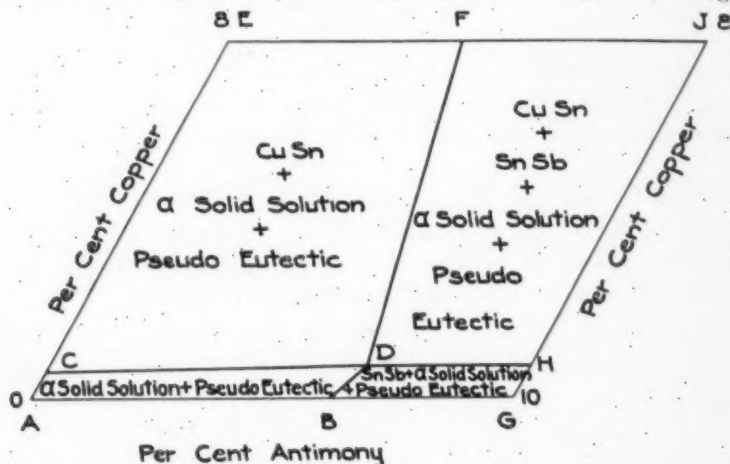
Fig. 1—Tin-Antimony-Copper Constitution Diagram.

white needles are of the intermetallic compound, CuSn. This was the first constituent of this alloy to freeze on cooling. The gray areas are sections of the alpha solid solution of the tin-antimony system; the black areas, containing here and there minute crystallites of the intermetallic compound CuSn, are sections of pseudo-eutectic.

Fig. 4 is the structure of an alloy containing approximately 86 per cent tin, 6 per cent antimony, and 8 per cent copper. The structure of this alloy is similar to that shown in Fig. 3. The proportion of pseudo-eutectic in this alloy is, however, so small as to make it difficult to see it at the magnification used in this photomicrograph. The proportion of pseudo-eutectic in these alloys decreases as their antimony and copper contents increase.

Fig. 5 is the structure of an alloy containing approximately 91.5 per cent tin, 8 per cent antimony, and 0.5 per cent copper. This may be considered as a representative structure of the alloys of group 3, which lie within the area BGHD of Fig. 1. The outstanding feature of this structure is the constituent SnSb, which is present in the alloy in the form of cubes, sections of which are clearly seen in the photomicrograph. This constituent was the first to separate from the melt on cooling. It is imbedded in a matrix consisting almost entirely of alpha solid solution. The proportion of pseudo-eutectic present in the alloy is so small that it is impossible to detect at the magnification used.

Fig. 6 is the structure of an alloy containing approximately 89.5 per

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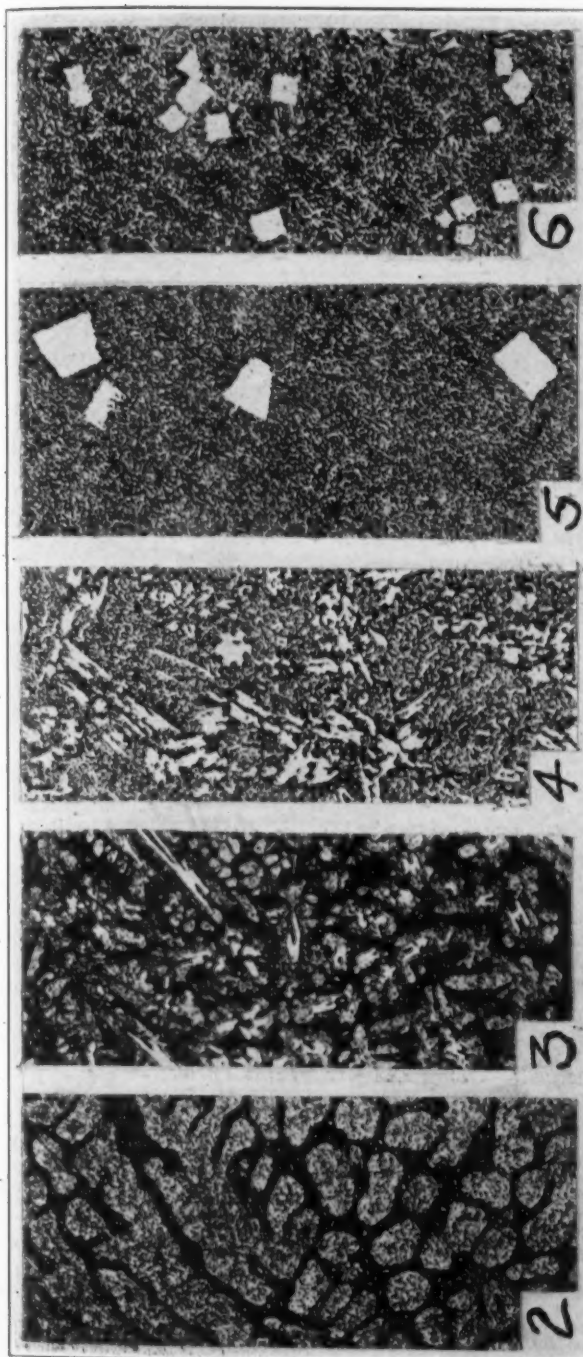
Fig. 2—Structure of an Alloy Containing Approximately 95.5 per cent Tin, 4 per cent Antimony, and 0.5 per cent Copper. × 150.
Fig. 3—Structure of an Alloy Containing Approximately 94 per cent Tin, 2 per cent Antimony, and 4 per cent Copper. × 150.
Fig. 4—Structure of an Alloy Containing Approximately 86 per cent Tin, 6 per cent Antimony, and 8 per cent Copper. × 150.
Fig. 5—Structure of an Alloy Containing Approximately 81.5 per cent Tin, 8 per cent Antimony, and 0.5 per cent Copper. × 150.
Fig. 6—Structure of an Alloy Containing Approximately 89.5 per cent Tin, 10 per cent Antimony, and 0.5 per cent Copper. × 150.

cent tin, 10 per cent antimony, and 0.5 per cent copper. This photomicrograph is a companion to Fig. 5 and is introduced for the purpose of emphasizing the apparent effect of the intermetallic compound, CuSn, on the size of the cubes of the tin-antimony compound, SnSb.
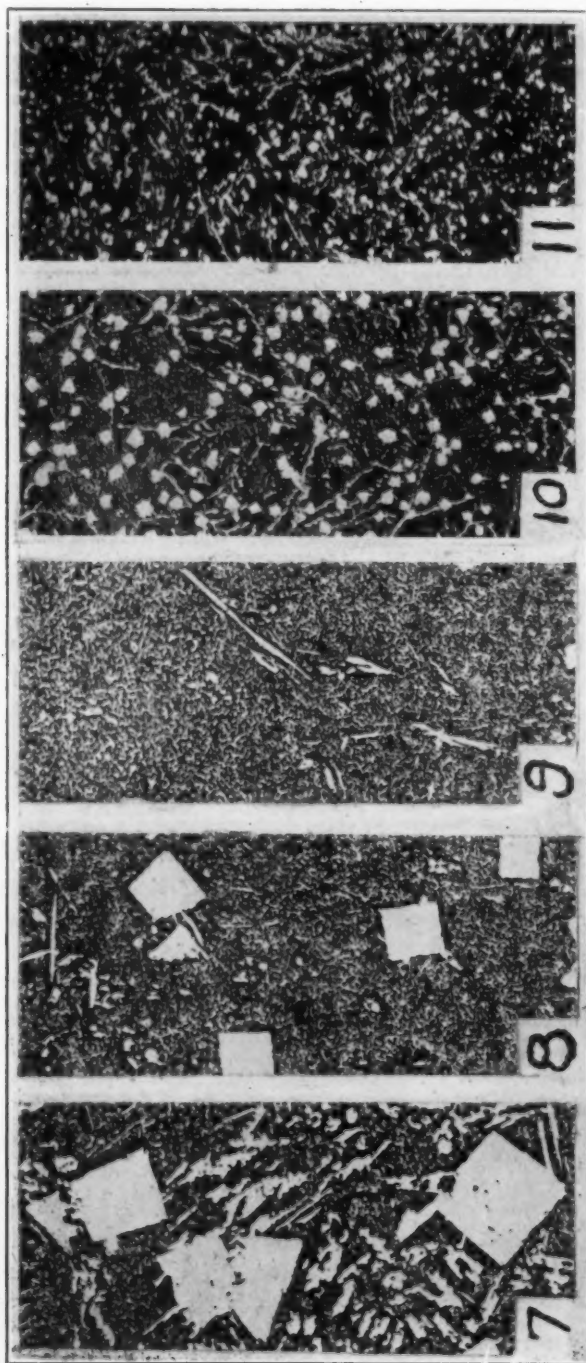
Fig. 7.—Structure of an Alloy Containing Approximately 82 per cent Tin, 10 per cent Antimony, and 8 per cent Copper. × 150.
Fig. 8.—Structure of an Alloy Containing Approximately 90 per cent Tin, 80 per cent Antimony, and 2 per cent Copper. × 150. This Represents the Average Structure at the Center of the Ingot.
Fig. 9.—Same Alloy as Fig. 8. This represents the Average Structure at the Edge of the Ingot. × 150.
Fig. 10.—Alloy Containing 89 per cent Tin, 8.7 per cent Antimony, and 2.3 per cent Copper. × 20. Pouring Temperature 500 Degrees Cent.; Mold Temperature 100 Degrees Cent. (Hudson and Darley[2]).
Fig. 11.—Same Alloy as in Fig. 10. × 20. Pouring Temperature 300 Degrees Cent.; Mold Temperature 100 Degrees Cent. (Hudson and Darley[2]).

Fig. 7 is the structure of an alloy containing approximately 82 per cent tin, 10 per cent antimony, and 8 per cent copper. This may be considered as a representative structure of the alloys of group 4, which lie within the area DHJF of Fig. 1. Here sections of needles of the inter-

metallic compound, CuSn, which was the first constituent to freeze on cooling, and of cubes of the intermetallic compound, SnSb, which was the second constituent to freeze on cooling, are found in a matrix of alpha solid solution. Some pseudo-eutectic is present in the structure but in such small amounts that it is difficult to detect at a magnification of 150 diameters.

It will be noted that the size of the cubes in Fig. 7 is considerably greater than that of the cubes in Fig. 6. This greater cube size is characteristic of the alloys of high copper content, other things apparently being equal. Ellis and Karelitz[2] are of opinion that "the difference in crystal size may be accounted for by assuming that the presence of primary CuSn facilitates the precipitation of SnSb and, as a result, a few large cubes are formed where, in the absence of the inoculating medium CuSn, a greater number of small cubes would be precipitated. The presence of needles of CuSn within the cubes of SnSb is a common occurrence in the alloys of high copper content."

Rapid cooling has the effect of suppressing the crystallization of SnSb in antimony-rich alloys in the neighborhood of the boundaries BD and DF of Fig. 1. It is therefore possible to obtain structures free from cubes of SnSb in certain alloys, if the same be cooled from the liquid state at a sufficient rate. This effect in an alloy, containing approximately 90 per cent tin, 8 per cent antimony, and 2 per cent copper, is brought out in Figs. 8 and 9, the former representing the average structure toward the center of an ingot of this alloy, the latter the average structure near the edge. It is possible that this effect might be of some importance in its bearing upon the anti-frictional behavior of the alloys lying near the boundaries BD and DF, and it may be well to emphasize the fact that the further removed an alloy is from these boundaries, the less likely is the normal crystallization of SnSb to be prevented.

In the case of alloys further removed from these boundaries the rate of cooling has an important effect on the dimensions of the cubes of SnSb, these decreasing in size as the rate of cooling is increased. The pouring temperature, except in so far as it affects the rate of cooling, is insignificant in its action in this regard.

In certain cases, however, the pouring temperature, has quite an important effect upon the manner in which the needles of CuSn distribute themselves. Hudson and Darley[3] have shown in the case of an alloy containing 89.0 per cent tin, 8.7 per cent antimony, and 2.3 per cent copper, poured at known temperatures into an iron ingot mold 1⅞ inches long by ⅞ inch wide by ½ inch deep, previously heated to varying degrees, that (a) in those cases where the pouring temperature was above about 300 degrees Cent., the primary needles of CuSn separate from solution in the form of long, feathery or chain-like crystallites, generally arranged in a very distinct angular fashion and frequently in clearly defined triangular groups (Fig. 10); and (b) in those cases where the pouring temperature

[2]Ellis and Karelitz, A. S. M. E. *Transactions,* 1928, MSP-50-11, p. 13.

[3]Hudson and Darley, *Journal* Institute of Metals, 1920, Vol. XXIV, p. 361.

Fig. 10—Alloy Containing 89 per cent Tin, 8.7 per cent Antimony, and 2.3 per cent Copper. X 20. Pouring Temperature 500 Degrees Cent.; Mold Temperature 100 Degrees Cent. (Hudson and Darley[2]).

Fig. 11—Same Alloy as in Fig. 10. X 20. Pouring Temperature 300 Degrees Cent.; Mold Temperature 100 Degrees Cent. (Hudson and Darley[2]).

mately 82 per s may be con- p 4, which lie of the inter-

was about 300 degrees Cent. or below, the CuSn, occurs chiefly as shorter and often much thicker needle-shaped crystals or in star-shaped forms. Well marked triangular groups were rarely seen and the general "grain size" was much smaller (Fig. 11).

Hudson and Darley suggest that for each alloy of this system there will exist a critical temperature below which pouring should always be conducted if the triangular groupings of the CuSn crystallites are to be avoided, it being assumed that triangular groupings are objectionable on account of their effect on the properties of the alloy. To what extent, however, the results of experiments on relatively heavy sections can be applied to the thin linings ($\frac{1}{16}$ to $\frac{1}{4}$ inch) of white metal so widely used in engineering work is open to question. In general, pouring temperatures of from 400 to 450 degrees Cent. do not appear to have a deleterious effect on the alloys of this system when cast in the form of thin linings.

CONSTITUTION OF THE TIN-ANTIMONY-COPPER ALLOYS USED AS BEARING METALS

By O. W. Ellis[*]

THE tin-antimony-copper system has been the subject of considerable investigation[1 to 8] but no work on the equilibrium of these alloys has yet been published. The structure of the alloys of this system which are of the greatest industrial importance (from 0 to 8 per cent of copper and from 0 to 10 per cent of antimony) are, however, but slightly affected by treatments designed to bring the alloys into a state of equilibrium. Hence, the following discussion, which is mainly concerned with the industrial alloys, would require but slight alteration to bring it into line with equilibrium requirements. In the case of alloys containing more than 8 per cent of copper, the structures of the alloys, as cast, differ appreciably from those of the alloys after suitable annealing; therefore, when alloys containing more than 8 per cent of copper are referred to, it must be remembered that only the alloys as cast are being considered.

The constituents present in these high tin ternary alloys as cast are four in number, as follows:

1. The alpha solid solution, containing tin and copper. In these alloys as cast this constituent is characterized by a cored structure which can be removed by annealing for sufficient time at 100 degrees Cent.

2. The intermetallic compound, CuSn, containing approximately 34.8 per cent copper. This constituent is present in these alloys in the form of needle-shaped crystallites which vary in size according to the rate at which the alloys are cooled. There seems to be no doubt that in these alloys, as cast, the needle shaped constituent is none other than the intermetallic compound CuSn.[9] When these alloys, however, are brought into a state of equilibrium by prolonged annealing this compound is replaced by a solid solution containing approximately 40 per cent copper.[10, 11]

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3. The intermetallic compound of copper and tin, Cu_3Sn, containing approximately 61.5 per cent copper. This compound occurs in these alloys in the form of needle shaped crystallites. It does not occur in alloys containing less than 8 per cent of copper and will not enter into the subsequent discussion of this system.

4. The intermetallic compound of tin and antimony, SnSb, containing approximately 50.4 per cent antimony, which appears in the microstructure of these alloys, as cast, in the form of very sharply defined cubes
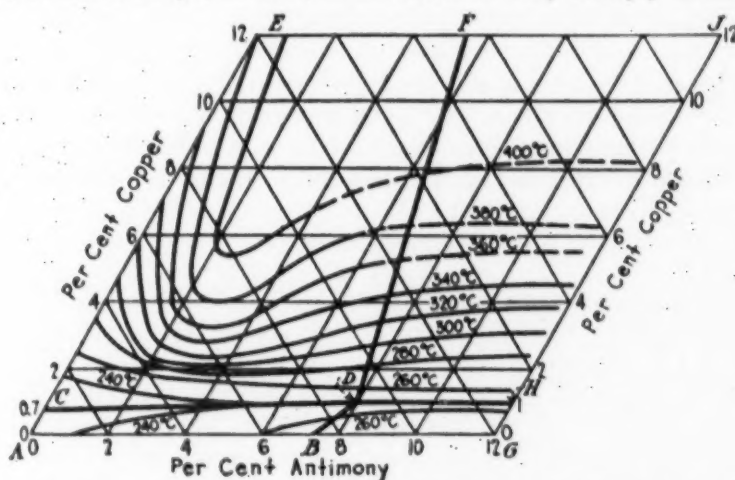
Fig. 2—Contours of Liquidus Surface of Space Model for Tin Corner of System.

and sections of cubes. Whether, under conditions of equilibrium, this compound would be replaced by the beta solid solution of the tin-antimony system is uncertain. The constitution of the latter system deserves much fuller investigation.

It is convenient to arrange the alloys in four groups in accordance with their microstructure, as shown in Fig. 1, page 472.

1. The first group of alloys, as cast, are characterized by a hypoeutectic structure, comprising alpha solid solution and a eutectic, or pseudo-eutectic, which is composed of the alpha solid solution and the intermetallic compound, CuSn, of the copper-tin system.

2. The alloys in the second group, as cast, contain three micro-constituents: the intermetallic compound, CuSn; the alpha solid solution; and the pseudo-eutectic already described.

3. The third group of alloys, as cast, comprise three micro-constituents: the intermetallic compound, SnSb; the alpha solid solution; and pseudo-eutectic. The proportion of pseudo-eutectic in these alloys is very small.

4. The fourth group of alloys, as cast, comprise four micro-constituents: intermetallic compound, CuSn; intermetallic compound, SnSb; alpha solid solution; and pseudo-eutectic, of which the proportion is very small.

These four groups lie within the areas outlined in Fig. 1. Group 1

occupies the area ABDC; group 2, the area CDFE; group 3, the area BGHD; and group 4, the area DHJF.

The contours of the liquidus surface of the tin corner of the space model for this system[3] are shown in Fig. 2. The solidus for all the alloys covered by this diagram is 225 degrees Cent. which is the eutectic temperature of the tin-copper system. In alloys which contain SnSb as a separate microconstituent, SnSb begins to crystallize on slow cooling at a temperature of 248 degrees Cent. Rapid cooling has the effect of suppressing the crystallization of SnSb at 248 degrees Cent. in antimony-rich alloys in the neighborhood of the boundaries BD and DF. It is possible that this effect might be of some importance in its bearing upon the antifrictional behavior of the alloys lying near the boundaries BD and DF, and it may be well to emphasize the fact that the further removed an alloy is from these boundaries, the less likely is the normal crystallization of SnSb to be prevented.

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[1]Smith and Humphries, Institute of Metals *Journal,* 1911, Vol. V, No. 1, p. 194.
[2]Campbell, American Society for Testing Materials *Proceedings,* 1913, Vol. 13, p. 630.
[3]Fry and Rosenhain, Institute of Metals *Journal,* 1919, Vol. XXII, p. 217.
[4]Thompson and Orme, Institute of Metals *Journal,* 1919, Vol. XXII, p. 203.
[5]Mahin and Broeker, Indiana Academy of Science *Proceedings,* 1919, p. 91.
[6]Hudson and Darley, Institute of Metals *Journal,* 1920, Vol. XXIV, p. 361.
[7]Bonsack, *Zeitschrift für Metallkunde,* 1927, Vol. XIX, p. 109.
[8]Ellis and Karelitz, American Society of Mechanical Engineers *Transactions,* 1928, MSP 50-11, p. 13.
[9]Isihara, Institute of Metals *Journal,* 1924, Vol. XXXI, p. 315.
[10]Shepherd and Blough, *Journal* Physical Chemistry, 1906, Vol. 10, p. 630.
[11]Haughton, Institute of Metals *Journal,* 1921, Vol. XXV, p. 309.

NICKEL AS AN ALLOYING ELEMENT IN STEEL

Thomas H. Wickenden[†]

Historical—The occurrence of nickel in meteorites and the working of this material into swords and tools undoubtedly accounts for the first use of nickel steel. Nickel, however, as an element was not recognized until isolated by Cronstedt in 1751. The first published report of its being alloyed with iron was by Stoddard and Faraday[1] in 1820. The next important development was the production started in 1885 by Marabeau of a wide series of iron-nickel alloys and nickel steel. These were studied and tested by James Riley of Scotland and the results were published in his now famous paper of 1889 on the "Alloys of Nickel and Steel." He was the first to demonstrate the general properties of nickel steels and pointed out their commercial value. During this period nickel had gradually become more available and cheaper because about this time nickel was produced in commercial quantities from both the New Caledonian and Canadian ores.

Riley's paper marked the beginning of active engineering and metallurgical interest in these steels. In 1891 the U. S. Navy conducted a competitive test between plain carbon and nickel steel armor plate. The nickel

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[1]*The Quarterly Journal of Science, Literature and the Arts,* 1820, Vol. IX, p. 324.

steel proved decisively superior and stimulated a world wide interest in this material for ordnance. In 1899 Hadfield[2] published an exhaustive paper on the "Alloys of Iron and Nickel." Since then it has been the subject of active interest and investigation by many scientists, and today finds an exceedingly wide field of industrial applications.

Alloying Nickel in Steel—Nickel is a white, malleable, magnetic metal harder and stronger than iron, and has a melting point of approximately 2625 degrees Fahr. Nickel is supplied to the steel maker in the form of shot or ingots of about 99 per cent purity, but the greatest demand is for electrolytic cathodes of still higher purity. As nickel is not oxidized in the bath, in fact nickel is chemically negative to the iron, it may be added at any time practically without loss. Part of the nickel is generally secured by charging nickel steel scrap; the balance may be added with the charge or at sufficient time before tapping to ensure proper diffusion. As nickel has no deoxidizing properties, it does not prevent blowholes, hence the steel to which it is added should be well made. Nickel checks segregation and tends to produce a fine grain primary crystal but is primarily added to steel for its beneficial effect on the physical properties.

The Effects of Nickel in Steel—Iron and nickel are soluble in all proportions in the molten state and remain as a solid solution on solidification. At normal temperatures nickel is a solid solution in the iron phase, either ferrite or austenite, depending on the composition. The equilibrium diagram of iron and nickel is given in another section of this Handbook. An examination of this shows some of the fundamental effects of adding nickel to iron. The addition of nickel reduces the temperature of the Ac_3 transformation and introduces a considerable spread between the temperature of beginning and the end point. The Ar_3 point is reduced to a still greater extent, in fact with 33 to 34 per cent of nickel it is not reached at liquid air temperatures.

The constitutional diagram of nickel steel as determined by Guillet is shown in Fig. 1. It indicates that after slow cooling, depending on the nickel and carbon content, nickel steel may exist as three separate microstructures: pearlite, martensite, and austenite.

The lines of demarcation are not as sharply defined as indicated by the diagram as the phases tend to overlap. For any given composition an increase in cooling rate will change the structure in the same direction as increasing the nickel or carbon content.

The general effect of increasing quantities of nickel on the mechanical properties of a cast nickel steel with 0.25 per cent carbon is illustrated in Fig. 2. The strength is gradually increased, reaching a maximum of 180,000 lbs. per sq. in. with 12 per cent of nickel. Data on carbonless nickel-iron alloys, annealed, by Burgess and Ashton show the same maximum strength, but at 18 per cent of nickel. Carbon has the effect of shifting the diagram to the left and accentuating the martensitic stage. It should be noted from the diagram that the abrupt changes in mechan-

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[2]*Proceedings* Institute of Chemical Engineers, Vol. CXXXVIII.

ical properties correspond to the changes in the micro-constituents of the
steel.

Classification of Steel—Characteristics of the three groups of steel will
be discussed in detail.

The Pearlitic Steels—The pearlitic group, steels containing from 0.5 to
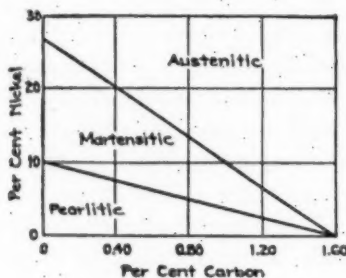
Fig. 1—Constitutional Diagram
of Nickel Steel (Guillet).

6.0 per cent nickel, are most widely used commercially. The principal ef-
fect of nickel on this group of steels is as follows:

1.   Nickel depresses the critical point and lowers the eutectoid ratio.
The effect of nickel in lowering the Ac critical range is well illustrated
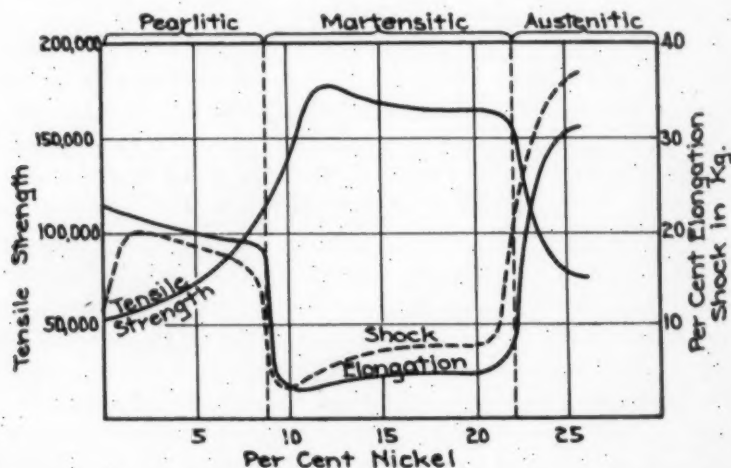
Fig. 2—Comparative Physical Properties of Nickel Steels with
0.25 Per Cent Carbon. Not Heat Treated (Bullens).

by Scott[3] in Fig. 3. He further states that the Ac_1 is lowered 18.9 degrees
Fahr., the Ar_1, 38.7 degrees Fahr., and the eutectoid ratio is decreased ap-
proximately 0.042 per cent carbon for each per cent of nickel added. Thus,
with a normal eutectoid ratio of 0.90 per cent carbon in plain carbon steel
this is reduced to approximately 0.75 per cent for 3.5 per cent nickel and
0.70 per cent for 5.0 per cent nickel steel. The lower critical range allows
the use of lower quenching temperature when heat treating nickel steels,

[3]*Bureau of Standards Bulletin* No. 376, p. 212. Critical Ranges of Some Commercial
Nickel Steels.

and consequently results in less scaling, warping, and cracking of the part. This also permits a saving of fuel and less wear on the heat treating equipment.

2. Nickel increases the strength, yield point, and hardness without a corresponding loss of ductility in both the annealed and heat treated
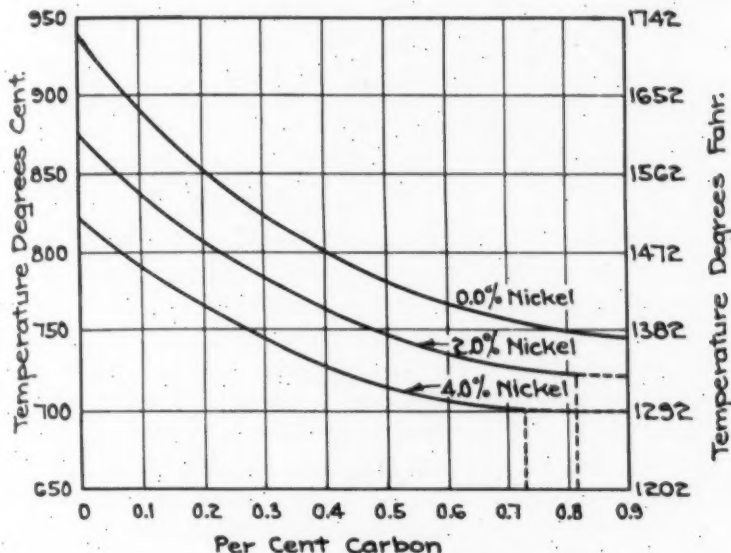
Fig. 3—Effect of Nickel on the Ac Critical Range.

steels. The following table gives data on an annealed 0.30 per cent carbon steel with different nickel contents.

Carbon	Nickel	Tensile Strength Lbs. Per Sq. In.	Yield Point	Elongation in 2 Inches Per Cent	Reduction of Area Per Cent	Brinell Hardness Number
0.30	0	70,000	40,000	26.0	50.0	145
0.30	3.50	90,000	60,000	27.0	55.0	175
0.30	5.00	95,000	65,000	28.0	56.0	185

These characteristics of nickel are utilized in rolled structural shapes, boiler plates, and large forgings or castings which, due to size or shape, are impractical to heat treat beyond normalizing and annealing.

Abbott,[4] after a great many tests, observed the average effect of nickel up to 8 per cent, for fully annealed steel, to be as follows:

1.0 per cent of nickel increased the elastic limit......................4,000 lbs. per sq. in.
1.0 per cent of nickel increased the tensile strength....................4,200 lbs. per sq. in.
1.0 per cent of nickel increased the reduction of area........................0.5 per cent
1.0 per cent of nickel decreased the elongation................................1.0 per cent

If annealed steels of the same strength are compared, the advantages of nickel in increasing the ductility are very evident from the following results:

[4]Rolled Nickel Steel, American Society for Testing Materials *Proceedings*, Vol. 17, Part 2, 1917, p. 11.

Carbon	Nickel	Tensile Strength Lbs. Per Sq. In.	Yield Point	Elongation in 2 Inches Per Cent	Reduction of Area Per Cent
0.40	0	85,000	42,000	25.0	45.0
0.20	3.50	85,000	55,000	32.0	65.0

In heat treated steels the benefit of nickel is still more evident by the following:

Carbon	Nickel	Tensile Strength Lbs. Per Sq. In.	Yield Point	Elongation in 2 Inches Per Cent	Reduction of Area Per Cent	Brinell Hardness Number
0.38	0	110,000	81,000	19.0	46.0	223
0.37	3.27	110,000	90,000	28.0	64.7	217

3. Nickel slows down the critical rate of hardening. This property affects its use in several ways. Oil can be used as a quenching medium which reduces the danger of warping or cracking on intricate shapes. Parts of large size can be hardened to considerable depth. Slow cooling tends to produce a fine grain pearlite tending toward sorbite.

4. Nickel retards the rate of grain growth at elevated temperatures. Nickel steels may be heated for long periods of time at elevated temperatures without serious injury. Thus it minimizes the danger of overheating during heat treatment. This feature, combined with the lower critical points, makes the nickel steels excellent for casehardening. As compared to carbon steel, they also show greater uniformity of carbon penetration. They are amiable to a single quenching treatment, develop greater core strength, and the case is tougher as shown by a less tendency of the nickel steels to show grinding cracks. The case will develop file-hardness on oil quenching.

5. Nickel increases the resistance to fatigue endurance ratio. McAdam[5] states the addition of nickel to annealed carbon steel apparently strengthens the ferrite with the result that the endurance ratio is raised. The full effect of nickel on the endurance ratio, however, is obtained only by quenching and tempering.

6. Nickel increases the resistance of steel to corrosion. This is easily demonstrated by the resistance to etching shown by 5.0 per cent nickel steels. This property is made use of by adding nickel to boiler tubes for use with bad water.

7. Nickel improves the impact properties of steel as shown by the following values[6] from a 0.45 per cent plain carbon steel and 0.36 per cent carbon, 3.65 per cent nickel steel:

			Izod Foot Pounds	
			0.45% C	0.36% C, 3.65% Ni
Normalized				
Tensile Strength	100,000 Lbs. Per Sq. In.		31	36
Heat Treated				
Tensile Strength	115,000 Lbs. Per Sq. In.		48	72
Tensile Strength	125,000 Lbs. Per Sq. In.		32	65
Tensile Strength	150,000 Lbs. Per Sq. In.		15	36

[5]D. J. McAdam, Jr. "On Endurance Properties of Steel," American Society for Testing Materials *Proceedings*, 1923, Vol. 23, Part 2, p. 99.

[6]Auto. Steel Research Report, 1920.

This shows at equal tensile strengths the nickel steel is tougher and has greater resistance to shock, especially as the tensile strength is increased.

8. Nickel is used with other alloying elements in pearlitic steel. The principal alloying elements used with nickel are chromium, molybdenum, vanadium, silicon, and manganese. As nickel strengthens the ferrite it makes an ideal base for the addition of alloys of the carbide forming group.

The nickel-chromium steels are produced with a wide variety of compositions. The analyses, properties, and further data on the applications and heat treatment of these alloys are given in other articles of this Handbook and the S.A.E. Handbook.

Nickel-molybdenum makes an excellent carburizing steel and nickel-
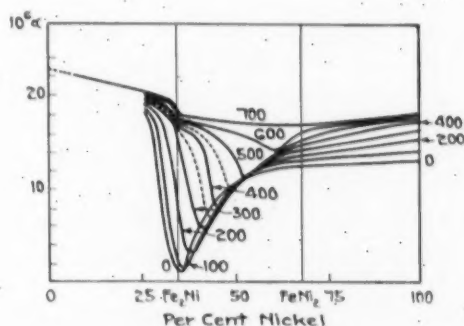
Fig. 4—Variation in True Coefficient of Linear Expansion of Nickel Steels at Various Temperatures (Chevenard).

chromium-molybdenum an excellent steel for heavy sections requiring high strength and good impact properties.

Nickel-silicon and nickel-silicon-molybdenum steels were developed during the late war period for light armor plate. They show unusual combinations of properties at high strengths.

Martensitic Steels—This group with 10 to 22 per cent of nickel is not much used. The steels are hard, almost impossible to machine; certain combinations are brittle, and give very little response to heat treatment.

Many corrosion resistant steels will be found with nickel within this range but it is usually in combination with other elements which tend to make the steel austenitic.

Austenitic Steels—The austenitic range of nickel steels presents a most fascinating study as their properties are so varied that they constantly unfold new and interesting engineering and scientific applications.

The straight iron-nickel group has a variety of uses as follows:

25-30 Per Cent Nickel—Alloys with this nickel range are used for corrosion resistance and for low temperature electrical resistance wire.

20-30 Per Cent Nickel—Steels with this nickel range are non-magnetic after cooling at normal rates from forging or rolling temperatures to atmospheric temperature. Their magnetic condition can be restored by cooling to liquid air temperature.

30-40 Per Cent Nickel—This nickel range presents an interesting anomaly in coefficient of expansion. As the nickel content is increased above 30 per cent, the coefficient of linear expansion is rapidly lowered until it reaches a minimum of nearly zero coefficient between 35 and 36 per cent of nickel. This phenomenon was discovered by Guillaume, and to carefully prepared and heat treated alloys of this composition he gave the name of "Invar." In its finer state it is used for the production of fine instruments, standard bars, measuring tapes, clock pendulums, and balance wheels of watches. In a more commercial grade with wider range of nickel this alloy is finding use as struts in aluminum pistons to control the overall expansion to that of cast iron. At elevated temperatures the point of minimum expansion is lower with higher nickel ratios as is shown in Fig. 4.

Elinvar—Elinvar is an alloy having a nonvariant elastic modulus with temperature change. It was developed by Guillaume and contains 36 per cent nickel and 12 per cent chromium. Its principal application is in compensating springs for high grade watches and chronometers.

Plantinite—Plantinite is a nickel steel with about 46 per cent nickel which has the same coefficient as platinum.

50 Per Cent Nickel—An alloy of 50 per cent nickel and iron is used for its magnetic properties, having a high permeability at low field strengths. It is finding application in the radio field.

78.5 Per Cent Nickel—This is another magnetic nickel-iron alloy with high permeability at low field strengths; also the magnetic hysteresis loss is very low. It is finding use in telephone equipment and as loading tape in submarine cables. It is known as Permalloy.

Corrosion and Heat Resisting Alloys—The recent engineering interest in corrosion, heat resisting, and high temperature materials has led to the development of a large number of new alloys, a great many of which contain nickel. Some typical compositions are:

Nickel	Chromium	Silicon	Other Elements		Iron
80.0	20.0	0.75			Balance
60.0	13.0	0.75			Balance
58.0	12.0	...	Tungsten	4.0	Balance
57.0	...	...	Molybdenum	20.0	Balance
24.0–26.0	17.0–19.0	2.50–3.00			
20.0–22.0	7.0– 8.0	0.80–1.75			
20.0–22.0	25.0–27.0	2.00–3.00			
10.0–12.0	20.0–22.0	1.00–3.00			
7.0–10.0	17.0–20.0	1.00–3.00			
9.0	12.0	3.00	Tungsten	6.0	
7.0– 9.0	11.5–14.0	2.00–3.00			
1.5	10.0–11.0	1.00	Aluminum	2.0	Balance

Most of these alloys are not affected appreciably by heat treatment. The wide range of analyses has been developed to meet corrosion resistance in various media, oxidation resistance at high temperatures, and high strength at elevated temperatures. Many of the steels with nickel around 20 per cent are non-magnetic at ordinary temperatures.

THE ENGINEERING INDEX

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In the preparation of the Engineering Index by the staff of the American Society of Mechanical Engineers some 1700 domestic and foreign technical publications received by the Engineering Societies Library (New York) are regularly searched for articles giving the results of the world's most recent engineering and scientific research, thought, and experience. From this wealth of material the A. S. S. T. is supplied with this selective index to those articles which deal particularly with steel treating and related subjects.

ALLOY STEEL

Notes on Alloy Steels. J. H. Andrew. *Iron and Coal Trades Rev. (Lond.)*, vol. 119, no. 3222, Nov. 29, 1929, pp. 833 and 840.

Practice has shown that smaller dendritic pattern in cast indicates better ingot; relation between dendritic structure and mechanical properties, shown by results of test; advantages associated with hot work; constitution and mechanical strength; formation of carbides; choice of alloy steels; depth hardness; temper; brittleness; quenching temperatures; influence of rate of cooling on fracture. Abstract of paper read before Instn. Engrs. and Shipbuilders in Scotland.

Modern Trends in Alloy Steels. *Iron Age*, vol. 124, no. 24, Dec. 1929, pp. 1593-1595.

Review of meeting of American Institute of Mining and Metallurgical Engineers at which James Douglas Medal was awarded to P. D. Merica, with brief abstracts of papers presented; Recent Developments in Applications of Steels, C. F. MacQuigg; tests on chromium steel rails; stronger steels for airplanes; improvements made in cutting tools; distribution of 20,000,000 lb. of high-speed steel produced annually; about 25,000 tons of high-chromium steel made each year.

ANTI-CORROSIVE. Resistance of Alloy Steels to Corrosion (Les alliages ferreux résistant aux corrosion), A. Matagrin. *Nature (Paris)*, no. 2819, Oct. 15, 1929, pp. 354-360, 8 figs.

Development and properties of nickel steel and refractory alloys, cast iron with silicon, and steel with chromium are discussed.

ALLOYS

ANTI-CORROSIVE. Experience Dictates Construction Materials for Handling Corrosive Agents, A. H. Cooper. *Chem. and Met. Eng.*, vol. 36, no. 12, Dec. 1929, pp. 747-750.

Review of literature showing present practice in handling sulphuric, nitric, hydrochloric and acetic acids, alkalis, ammonia and hydrogen sulphide.

Properties of Corrosion-Resisting Alloys, P. S. Menough. *Blast Furnace and Steel Plant*, vol. 17, no. 12, Dec. 1929, pp. 1810-1815, 13 figs. See also *Heat Treating and Forging*, vol. 15, no. 12, Dec. 1929, pp. 1567-1572.

Work hardening, welding, and machining, of nickel-chromium steels of various compositions; new high-temperature high-strength alloys; foundry practice in chromium and chromium-nickel steels; liquid shrinkage; solid shrinkage; troubles with heavy and thin walled sections adjoining.

Properties of Corrosion-Resisting Alloys, P. S. Menough. *Blast Furnace and Steel Plant*, vol. 17, no. 11, Nov. 1929, pp. 1648-1651 and 1655, 3 figs.; see also *Heat Treating and Forging*, vol. 15, no. 11, Nov. 1929, pp. 1437-1440, and 1450, 3 figs.

Selection of compositions and methods in fabricating parts involves familiarity with technique that is gained only through experience; effect of adding chromium to steel up to stainless analysis and beyond; influence of carbon; some weaknesses of high-chromium steels; resistance to high temperature and corrosion; welding; machining; abrasion resistance. (To be continued.)

ELASTICITY. On the Measurement of the Elastic Constant, Lattice Constant, and Density of Binary Alloys in the Range of Solid Solution, Z. Nishiyama. *Tohoku Imperial Univ.—Science Reports (Sendai)*, vol. 18, no. 3, Oct. 1929, pp. 359-400, 16 figs.

Author has measured Young's modulus of elasticity, lattice constant, and density of binary solid solutions; it was found that observed density agrees satisfactorily with that calculated from lattice constant; relation of elastic constant to density, to hardness or to atomic volume is discussed.

ALUMINUM ALLOYS

Aluminum Alloys. *Metallurgist (Supp. to Engineer, Lond.)*, Dec. 27, 1929, pp. 178-179.

While motor car might profit very much by saving of weight, there are other branches of engineering where possibilities of light alloys do not appear to have been fully appreciated; this applies most markedly to all forms of moving objects which have to be started and stopped many times, whether these be parts of fast-moving machinery or vehicles of tube train or suburban railway.

ALDREY. Aldrey, New Construction Material for Conductors of Electric Lines (El "Aldrey" nuevo material de construccion para conductores de lineas aéreas), F. Guhrauer. *Energia Electrica (Madrid)*, vol. 31, no. 20, Oct. 25, 1929, pp. 235-239, 2 figs.

Description of aluminum alloys with magnesium, silica, and iron; comparison of properties of copper, pure aluminum, aluminum with steel core, and aldrey; examples of successful use of aldrey for electric transmission lines.

CORROSION. Corrosion of Aluminum Alloys in Superheated Steam (La corrosion des alliages d'aluminium dans la vapeur d'eau surchaufee), L. Guillet and Ballay. *Académie des Sciences—Comptes Rendus des Séances (Paris)*, vol. 189, no. 16, Oct. 14, 1929, pp. 551-553.

Deterioration of aluminum alloys in superheated steam at 300 to 350 deg. is due to formation of alumina, and occurs at junction of grains; it is influenced by presence of other metals; and is greatest for purest samples of aluminum in both still and circulating steam, and least for alpax (13 per cent Si); corrosion also appears to depend on thermal and mechanical history of alloy.

FORGING. Forge Shop for Aluminum Alloys. *Iron Age*, vol. 124, no. 23, Dec. 5, 1929, pp. 1521-1522.

Methods employed by Aluminum Co. of America, for forging aluminum are described; furnaces, hammers and heat-treating equipment installed for daily capacity of ten tons; sizes range from tiny parts to airplane propellers or locomotive side rods; surface of forging billets must be perfect; hammers operated by 100-lb. air; forge shop of glass and steel; heat treating done in pit furnaces; rigorous testing and inspection.

GAS REMOVAL. Removal of Gases from Aluminum Alloys, Claus. *Metallurgist (Supp. to Engineer, Lond.)*, Dec. 27, 1929, pp. 190-191.

Methods for removing dissolved gases from molten aluminum alloys are described and results given of experiments showing effect of method developed by author and Kalaehne applied to Y-alloy and to 8 per cent copper alloy. Review of paper in Zeit. fuer Metallkunde, Aug. 1929.

ALUMINUM-COPPER ALLOYS

Copper-Aluminum Alloys of Manganese, Tin and Cobalt (Sur les cupro-aluminium au manganése à l' étain et au cobalt), E. Morlet. *Académie des Sciences—Comptes Rendus (Paris)*, vol. 189, no. 2, July 8, 1929, pp. 102-104.

Resistance and Brinell hardness of Cu-Al alloys are increased by addition of manganese; tin makes alloys fragile and forms small cavities.

TERNARY. Study of Cupro-Aluminum Alloyed with Manganese, Tin or Cobalt (Etude sur les Cupro-Aluminium au manganèse, à l'étain et au cobalt). *Revue de Métallurgie (Paris)*, vol. 26, nos. 9 and 10, Sept. and Oct. 1929, pp. 464-487 and 554-569, 68 figs.

Account of detailed study of ternary alloys of aluminum and copper; preparation of alloys, methods, thermal and dilatometric analyses, and results obtained; influence of heat treatment on properties; micrographic analyses.

ALUMINUM METALLOGRAPHY

Slip-Bands of Compressed Aluminum Crystals, K. Yamaguchi. *Inst. of Physical and Chem. Research—Sci. Papers (Tokyo)*,

vol. 11, no. 205, Sept. 8, 1929, pp. 223-241, 11 figs. partly on supp. plates.

Preparation of compression test pieces and determination of their orientations; analysis of distortion when specimen slips on one plane; tentative theory of work-hardening of metal; tentative theory about mechanism of self-stopping of slip is proposed.

BEARING METALS

Bearing Alloys, R. T. Rolfe. *Mech. World (Manchester)*, vol. 86, nos. 2237 and 2238, Nov. 15 and 22, 1929, pp. 462-463 and 484-486, 2 figs.

Properties and composition of bearing metals used in commercial practice; typical tin-base alloys; die-casting alloys of tin group; white metals used in German Navy; Nov. 22. Variations of structure in cooling; adhesion of joint; special methods in lining. (Concluded.)

LEAD BASE. Nickel in Lead-Rich Bearing Metals (Nickel in hoch bleihaltigen Lagermetallen), C. L. Ackermann. *Zeit. fuer Metallkunde (Berlin)*, vol. 21, no. 10, Oct. 1929, pp. 339-341, 7 figs.

Influence of nickel on strength of lead bearing metal; improvement of gliding properties; nickel in presence of small copper additions; discussion of some commercial alloys refined with nickel. (To be concluded.)

CASE HARDENING

Cause and Remedy of Defects in Case-hardening Work, H. W. Keeble. *Iron and Steel Industry (Lond.)*, vol. 3, no. 2, Nov. 1929, pp. 39-40 and 42.

Outstanding troubles with which author has met while connected with heat treatment shop are discussed; soft spots; variation in depth of case; exfoliation; causes of cracking; hair cracks and quenching; distortion.

CARBURIZING. Abnormality in Case-Carburized Steel, C. H. Herty. *Min. and Met. Investigations Bul.*, no. 45, 1929, 66 pp., 78 figs.

Discussion of work involving four years of investigation; review of previous work on abnormality of steels; present investigation shows that certain factors promote structural normality, while others promote structural abnormality; degree of normality, or abnormality of given steel will depend upon balance between these factors; enumeration of factors that promote structural abnormality. Bibliography.

Relation of Time and Temperature to Depth of Case in Carburizing of Carbon Steel, J. Sorenson. *Fuels and Furnaces*, vol. 8, no. 1, Jan. 1930, pp. 41-43.

Depth of case obtained in definite length of time at definite temperature will vary with nature of work, chemical composition of steel, carburizing agent, temperature used and length of time held at carburizing temperature.

CASTINGS

X-RAY ANALYSIS. X-Ray Analysis in Foundry Practice (Roentgentechnik in der Geisserei), W. B. Bartels. *Giesserei-Zeitung (Duesseldorf)*, vol. 26, nos. 22 and 23, Nov. 15 and Dec. 1, 1929, pp. 632-634 and 666-672, 29 figs.

Notes of use and advantages of X-ray analysis of castings; considerations of cost;

details of chemical and physical crystallography; five fundamental possibilities of fine structural analysis are enumerated.

CAST IRON

ALLOY. Study of Some Cast Irons Containing Nickel and Copper (Recherches sur quelques fontes au nickel et au cuivre), M. Ballay. *Revue de Métallurgie (Paris)*, vol. 76, no. 10, Oct. 1929, pp. 538-553, 26 figs.

Cast iron containing nickel, copper, and chromium are of interest because of corrosion resistance in different solutions, their resistance to oxidation, and their heat resistance; they evidently cannot be compared to nickel-chromium and similar steels, but they are much cheaper and can replace economically brasses and bronzes in certain cases.

HIGH TEST. "Giree" Cast Iron (La fonte "Giree"), F. Girardet. *Société Industrielle de l'Est—Bul.* no. 193, Apr.-June, 1929, pp. 18-39, 14 figs.

Discussion of improved properties and structure secured in cast iron made in melting furnace with gyratory hearth; by jolting iron in liquid state iron is freed from gases and thoroughly mixed and graphite dissolved; results of corrosion experiments.

MOLYBDENUM CONTENT. Molybdenum in Cast Iron, E. K. Smith and H. C. Aufderhaar. *Iron Age*, vol. 124, no. 23, Dec. 5, 1929, pp. 1507-1509, 5 figs.

Value of molybdenum content in cast iron is discussed; strength and hardness increased much more rapidly than machinability decreased; best results achieved with 1.5 per cent molybdenum; physical properties and structure.

CAST IRON PROPERTIES

Some Aspects of Semi-Steel Cast-Iron, W. E. Dennison. *Iron and Steel Industry (Lond.)*, vol. 3, no. 2, Nov. 1929, pp. 35-38, 4 figs.

Discussion of range of cast irons in mixing of which indefinite percentage of steel scrap has been originally used; steel loses its identity; alternative to use of steel; steel and corrosion question; limits of semi-steel cast irons; mixture factors.

CHEMICAL EQUIPMENT

CHROMIUM STEEL — How Chrome Steels Serve the Nitrocellulose Manufacturer, T. McKnight. *Chem. and Met. Eng.*, vol. 36, no. 9, Sept. 1929, pp. 530-531, 6 figs.

Brief account of application of chromium steel by Hercules Powder Co., Wilmington, Del.

CHROMIUM-NICKEL STEEL — What Chrome-Nickel Steels Offer to Chemical Engineers, J. A. Mathews. *Chem. and Met. Eng.*, vol. 36, no. 9, Sept. 1929, pp. 528-529, 3 figs.

Suggested uses of chromium-nickel steel in manufacture of chemical equipment; results of investigation conducted by crucible Steel Co. of Am.; discussion of Patents.

IRON-CHROMIUM ALLOYS — Chromium Irons and Steels Exhibit Widely Varying Properties, W. M. Mitchell. *Chem. and Met. Eng.*, vol. 36, no. 9, Sept. 1929, pp. 532-534.

Effect of chromium when added to ordinary carbon steel; classification of iron-chromium alloys; wide variety of properties and application in chemical industry.

SILICON ALLOYS — Where High-Silicon Irons Serve Chemical Industry. *Chem. and Met. Eng.*, vol. 36, no. 9, Sept. 1929, pp. 541-542, 3 figs.

Characteristics of silicon alloys and application in chemical industry; influence of silicon content on chemical resistance and transverse strength of high silicon-iron alloys.

CHROMIUM-NICKEL STEEL

Nickel Finds Greatest Use in Structural Steel, T. W. Gibson. *Eng. and Min. Jl.*, vol. 128, no. 14, Oct. 5, 1929, p. 541.

Four chief uses of nickel are: as component of alloys, surface coating for other metals, chemical or catalytic reagent, and as pure metal; tabular data on uses of nickel as estimated by president of International Nickel Co. Brief abstract from Mineral Industry.

COPPER ALLOYS

Copper-Oxygen System (Ueber das System Kupfer-Sauerstoff), R. Vogel and W. Pocher. *Zeit. fuer Metallkunde (Berlin)*, vol. 21, no. 10, Oct. 1929, pp. 333-337, 10 figs.

Diagram of state of copper-copper oxide system is studied, with special regard to solubility of oxygen in copper; case of decomposition of copper-cuprous oxide at lower temperatures is also described.

P.-M.-G. Metal. *Foundry Trade Jl. (Lond.)*, vol. 41, no. 692, Nov. 21, 1929, pp. 375-376, 5 figs.

Description of P.-M.-G. metal which is substitute for gun-metal and phosphor bronze; melting practice being carried out in pit fires of usual type, conditions being exactly same as for gun-metal, except that less notice was taken of casting temperature; molding; mechanical tests; heat-treated P.-M.-G.; comparison of cast P.-M.-G. metal with Admiralty gun-metal.

COPPER ANNEALING

Anomalies of Annealing After Cold Hardening of Copper and Brass (Anomalies du recuit apres écrouissage du cuivre et des laitons), F. Eugene. *Société d'Encouragement pour l'Industrie Nationale—Bul. (Paris)*, vol. 128, no. 5, May 1929, pp. 361-379, 20 figs.

Study of anomaly described by P. Nicolau which affects not only hardness but fitness for swaging characterized by deflection before rupture in Erichsen test, and, in less degree, elongation; dilatometric tests effected by differential dilatometer of Chevenard show that phenomena of annealing copper and brass is sufficiently different.

CORROSION PREVENTION

Rust Preventive Films the Armor of Modern Industry, B. Jeffs. *Can. Machy. (Toronto)*, vol. 40, no. 19, Sept. 19, 1929, pp. 68-70, 5 figs.

Discussion of means that may be taken to prevent industrial corrosion during processes of manufacture, during storage, during transportation, and during use; care to be taken in selection of rust preventives, and methods of applying them are outlined.

CRYSTALLIZATION

Crystals and Strength of Alloys, A. Glazunov. *Foundry Trade Jl. (Lond.)*, vol. 41, no. 682, Sept. 12, 1929, p. 184.

Discussion following presentation of paper entitled Crystalline Grains in Castings, which was given before International Foundrymen's Congress; star-shaped crystals and bearing metals; false and real structure of wrought iron; strength with grain-size reduction. (Concluded.)

COPPER-NICKEL ALLOYS

ELECTRIC RESISTANCE. The Effect of Drawing on the Temperature Coefficient of the Electric Resistivity of Constantan, R.S.J. Spilsbury. *Jl. of Sci. Instrument (Lond.)*, vol. 6, no. 11, Nov. 1929, pp. 357-358.

Temperature coefficients of electrical resistivity of sample obtained by author have varied from positive value of 60 parts in 1,000,000 per deg. cen. to negative one of 80 parts in 1,000,000.

DIES

DIE CASTING—ALUMINUM. Advantages of Aluminum as a Foundry Material, G. Mortimer. *Can. Foundryman (Toronto)*, vol. 20, no. 11, Nov. 1929, pp. 12-14.

Effects of low density high crystallization, shrinkage, high solid contraction, and weakness at high temperature of aluminum alloys on die design; venting and gating; castings in of bushes, inserts, and magnetic pole pieces; dressing of dies.

FORGING. The Construction of Drop Forging Dies, G. A. Smart. *Heat Treating and Forging*, vol. 15, no. 12, Dec. 1929, pp. 1564-1566, 12 figs.

Design of locks and counter locks for drop forging dies is discussed; various types of forgings illustrate design of these features; forgings without draft; elimination of end shift. (Continuation of serial.)

DURALUMIN

ALLOY ADDITIONS. Recent Investigations of Influence of Iron, Silicon, and Manganese on Refinement of Duralumin (Neue Untersuchungen ueber den Einfluss von Fe, Si und Mn auf die Duralumin-Veredelung), K. L. Meissner. *Zeit. fuer Metallkunde (Berlin)*, vol. 21, no. 10, Oct. 1929, pp. 328-332, 3 figs.

Investigations of copper-magnesium-aluminum alloys with stable content of copper and magnesium and varying additions of iron, silicon, and manganese for purpose of explaining their influence on improvement of duralumin; it was found that addition of silicon did not contribute to refinement.

ELECTRIC FURNACES

HEAT TREATING. Applications of Electric Furnaces. *Elec. Times (Lond.)*, vol. 76, no. 1990, Dec. 12, 1929, pp. 1035-1037, 7 figs.

Electric furnaces for ferrous industry, reheating, sheet annealing, wire annealing, annealing of steel castings, malleableizing of iron, tempering, carburizing, tool steel hardening and for non-ferrous industry including heat treatment of brass, copper, German silver, etc., and manufacture of aluminum;

growth of demand for electric heat-treatment furnaces from 1926 to 1929 is shown in graph.

HEAT TREATING VS. NATURAL GAS. Economic Use of Electric Furnaces, W. S. Scott. *Iron Age*, vol, 124, no. 26, Dec. 26, 1929, pp. 1725-1728, 3 figs.

Analysis of production costs of electricity vs. natural gas in heat treatment of high-grade shovels in plant of Paldwin Tool Works, Parkersburg, W. Va.; record of heat-treated production, load and coke consumption for 9 months; labor cost; vessel maintenance.

INDUCTION. Induction Furnaces. *Engineer (Lond.)*, vol. 148, nos. 3858 and 3859, Dec. 20 and 27, 1929, pp. 652-654 and 680-682, 17 figs.

Furnaces are of two kinds, standard frequency induction and those which employ high-frequency currents; while former is limited by metallurgical considerations and problems of electrical design, high-frequency furnace removes these limitations and gives wider field to metallurgical research, and to melting of metals having high melting points; details of Ajax Northrup and Ajax Wyatt furnaces; review of research on core-less induction furnaces; examples of furnaces and auxiliary equipment are illustrated.

METALLURGICAL. Report of the Electric Heat Committee of the Association of Iron and Steel Electrical Engineers for the Year 1928-29, G. H. Schaeffer. *Iron and Steel Engr.*, vol. 6, no. 10, Oct. 1929, pp. 533-540, 15 figs.

Electromagnetic Hevi-Duty hardening furnaces; Homo circulating-air electric tempering furnaces; carburizing and normalizing; Pull-Thru type furnaces; pit-type electric furnace; rotary-hearth furnace; automatic dumping; pusher-type furnaces; car-type furnace; electric steel-melting furnace; high frequency furnace for foundry.

NITRIDING. An Exceptionally Large Nitriding Furnace Installation. *Fuels and Furnaces*, vol. 7, no. 12, Dec. 1929, pp. 1907-1910 and 1934.

Installation consists of three electrically heated nitriding furnaces; 600-kw. car-type, 60-kw. box-type, and 10.5-kw. cylindrical-type furnace; parts up to 22 ft. in length are handled.

STEEL MAKING. Materials and Heat Balances of Steel-Making Electric Furnaces, N. I. Dorogov. *Vestnik Metallopromishlennosti (Moscow)*, nos. 8/9, Aug./Sept. 1929, pp. 138-153, 7 figs.

Report on series of tests on Heroult 3.5 tons furnace, installed in 1911 at Makeyev plant, with special reference to effect of interruption in operation, expenditure of electrodes, power for various operations, etc. (In Russian.)

ELECTRIC HEATING ELEMENTS

Non-Metallic Resistance Electric Heating Elements. *Engineering (Lond.)*, vol. 128, no. 3336, Dec. 20, 1929, pp. 814-815, 3 figs.

Details of Globar element which is made in rods of various sizes, and resembles in appearance compound known commercially as carborundum; application of elements to electric furnace is shown.

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14-815, 3 figs.
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ELECTROLYTIC CORROSION

Electrolytic Corrosion Due to Current Leakages and Its Prevention (Elektrolytische Press-Schaeden durch Stromlecken und vagabundierende Stroeme und deren Verhueting). G. Angel and Beck-Friis. *Chemiker-Zeitung (Koethen)*, vol. 53, nos. 57 and 59, July 17 and July 24, 1929, pp. 553-554, and 574-575, 6 figs.

Examples of electrolytic corrosion due to leakage of current in generating station and in neighborhood of underground conductors are described, together with methods for measuring resistance of local cells and loss of energy due to leakage; several methods of preventing corrosion, of electric railroad rails are mentioned.

FORGING, UPSET

Development of Upset Forgings, W. E. Crocombe. *Iron Age*, vol. 124, no. 23, Dec. 5, 1929, pp. 1513-1515, 2 figs.

Many advantages of upset method of forging are discussed and its development traced; saving one-fifth in weight and great improvement in output; uniform grain conditions give higher strength; much greater production obtained; upset method eliminates tong holds; plea for adequate cost analysis.

FURNACES

ANNEALING—NORMALIZING. Modern Furnaces for Normalizing.—The "Kathner" Furnace. *Iron and Coal Trades Rev. (Lond.)*, vol. 119, no. 3223, Dec. 6, 1929, pp. 865-867.

Description of principle and of advantages of single-sheet or thin-pack annealing over box annealing; description of Kathner furnace; mechanism of furnace; efficiency of insulation; application to black annealing tinplate.

FORGING. Forging and Normalizing Furnaces of Unusual Design Used at Ford Plant, J. B. Nealy. *Fuels and Furnaces*, vol. 7, no. 12, Dec. 1929, pp. 1859-1862, 1 fig.

Description of advancements in design and various unique features embodied in forging and normalizing furnaces at Fordson plant.

GAS. Gas Modernizes the Forge Furnace, J. B. Nealey. *Am. Gas Assn. Monthly*, vol. 11, no. 12, Dec. 1929, pp. 727-730, 4 figs.

Results of research by American Gas Association to ascertain cause of burning and its remedy together with study of conditions affecting burning and underlying causes of this behavior of steel; effect of oxidizing and reducing atmospheres on steels heated for forging was studied.

HEAT TREATING—FUELS. Fuels for Heat Treating. R. M. Keeney. *Iron Age*, vol. 125, no. 1, Jan. 2, 1930, pp. 58-59, 1 fig.

Trend in heat-treating fuel is toward refined sources; two important developments in 1929; gas and electricity expand despite lower cost of oil and coal; electricity shortens malleable annealing; gas replaces wood for annealing brass sheets; trend in 1930.

MELTING—PULVERIZED COAL. Uses Rotary Foundry Melting Furnace.

Iron Age, vol. 124, no. 24, Dec. 12, 1929, p. 1586.

Description of furnace of unique design called Brackelsburg which has recently been put into operation in Germany in production of high-quality gray and malleable castings; furnace is horizontal cylindrical shell, rotating during melting and superheating and fired with powdered coal; rate of revolution slow during melting. From Publication no. 131 of Kaiser Wilhelm Institute fuer Eisenforschung.

GLASS PLANTS

HEAT RESISTING STEEL. Heat-Resisting Steels with Special Reference to Their Application in the Glass Industry, R. J. Sarjant. *Soc. Glass Technology—Jl. (Lond.)*, vol. 13, no. 50, June 1929, pp. 167-182, 6 figs.

Object of paper is to give brief review of important metallurgical advance of recent years in domain of high-temperature engineering, and to indicate those properties of new materials, generally known as heat resisting, which are of special interest to glass industry.

HARDNESS TESTING

The Hardness Test in Practice, M. Moser. *Metallurgist (Supp. to Engineer, Lond.)*, Dec. 27, 1929, pp. 189-190.

Survey of modern methods of hardness testing; limitations of Brinell method are set forth and different Brinell testing machines described; Herbert pendulum hardness according to author's opinion, is still divided on its utility and field of application. Translated abstract of article previously indexed from *Kruppsche Monatshefte*, June 1929.

ROCKWELL. Mass Testing of Workpieces with the Rockwell Hardness Tester (Massenkontrolle von Werkstuecken), A. Rueter. *Maschinenbau (Berlin)*, vol. 8, no. 23, Dec. 5, 1929, pp. 790-794, 8 figs.

Rockwell hardness tester, its performance and operation, are described; on basis of actual experience use of instrument for mass testing is outlined, and influences leading to deviating indications are investigated; by means of frequency curves, tolerances for test pieces are determined.

INGOT MOLDS

DESIGN. Designing Ingot Molds to Reduce Rolling Mill Losses. *Rolling Mill Jl.*, vol. 3, no. 12, Dec. 1929, pp. 505-509, 7 figs.

Ingot molds design principles; discussion of both old and improved types of contours.

SEGREGATION. Reversed Ingot Segregation (Ueber umgekehrte Blockseigerung), W. Fraenkel and W. Goedecke. *Zeit. fuer Metallkunde (Berlin)*, vol. 21, no. 10, Oct. 1929, pp. 322-324, 4 figs.

In authors' opinion most important result of their investigation is fact that composition of crystals which are first solidified shows no reversed (inverse) segregation; in addition they discovered that layers were found in cast ingots directly beneath piping, composition of which is similar to that of crystals which are first separated.

IRON

ANCIENT. Ancient Iron and the Modern Science of Metals, V. N. Krivobok.

Fuels and Furnaces, vol. 7, no. 12, Dec. 1929, pp. 1839-1842, 10 figs.

Examination of pieces of ancient iron in Delhi column, foundation of Faruda Dhvaja column, old armor and weapons, and study of certain ancient manuscripts show that such distinct varieties of iron as iron, steel and pig iron were known in ancient times; discussion of results obtained by heat treatment of this ancient iron and results obtained with more modern heat-treating processes.

OXYGEN SOLUBILITY. Solubility of Oxygen in Solid Iron (Loeslichkeit des Sauerstoffs in festem Eisen), W. Bischof. *Stahl und Eisen (Duesseldorf)*, vol. 50, no. 1, Jan. 2, 1930, pp. 18-19.

Reference is made to work of R. Schenck and collaborators in Zeit. fuer anorganische Chemie, 1929, p. 97, on iron-oxygen system and influence of impurities of iron; work of W. Krings and J. Kempgens in same journal is also referred to; although these two works have contributed values which are in much better agreement with general theories of solubility of oxygen, it is claimed that problem is not yet satisfactorily solved.

IRON ALLOYS

TERNARY. Sulphur-Iron-Carbon System (Beitrag zur Kenntnis des Systems Schwefel-Eisen-Kohlenstoff), H. Hanemann and A. Schildkoetter. *Archiv fuer das Eisenhuettenwesen (Duesseldorf)*, vol. 3, no. 6, Dec. 1929, pp. 427-435, 25 figs.

Review of literature; determination of concentration and solidification temperature of ternary eutectoid; solidification phenomena in most important fields of ternary diagram iron-iron carbide-iron sulphide; method is described of determining carbon in presence of large quantities of sulphur.

IRON AND STEEL

BRITTLENESS. Brittleness of Homogeneous Iron (La fragilita nel ferro omogeneo) Sirovich. *Atti del Sindacato Provinciale Fascista Ingegneri di Milano (Milan)*, vol. 7, no. 9, Sept. 1929, pp. 289-296, 21 figs.

General discussion of commercial and psychological objections to adoption of homogeous iron of high strength; native and acquired brittleness; causes of brittleness; metallographic studies of brittleness; brittleness due to aging.

CRYSTALLIZATION. Crystal Structures of Iron and Steel, L. M. Jordan. *Heat Treating and Forging*, vol. 15, no. 11, Nov. 1929, pp. 1441-1443.

Investigation of individual iron crystals; strain and annealing methods in producing large crystals; relation of oxygen content; study of properties of iron single crystals; work hardening of hardened steels and its basis in crystalline modification.

IRON AND STEEL METALLURGY

Advancement in Iron and Steel Metallurgy During 1929, H. M. Boylston. *Fuels and Furnaces*, vol. 8, no. 1, Jan. 1930, pp. 23-36.

Discussion of progress made in ferrous metallurgy; development of stainless iron and steel; changes in dimensions and capacity of blast furnaces; increase in size of open hearth furnaces and developments in open hearth process; advancements in steel

making, uses of alloy steels and knowledge of their compositions; studies of constitutional diagrams, use of x-ray, and developments of automatic metallographic polishing machine; developments of various heat treatments and resulting structure of materials; advances made in nitriding process.

LOCOMOTIVE FIREBOXES

COPPER WELDING. Copper Welding (Die Kupferschweissung), R. Samesreuther. *V.D.I. Zeit. (Berlin)*, vol. 73, no. 49, Dec. 7, 1929, pp. 1731-1732, 4 figs.

Discussion deals principally with copper welding of locomotive fireboxes; comparison of riveted and welded copper fireboxes; method of welding and after-treatment of welds.

LOCOMOTIVE MANUFACTURE

ALLOY STEEL. Alloy Steels in Locomotive Construction, W. M. Barr. *Ry. Mech. Engr.*, vol. 103, no. 12, Dec. 1929, pp. 739-742 and 747, 6 figs.

Discussion of economic advantages, necessary precautions and advisability of joint research; lubrication of journal and crankpin bearings presents problem; alloy steel bolts, boiler plates, flues and springs; alloy steel plates tested in high-pressure boilers; use of nickel promotes uniformity in castings; photomicrographic illustrations showing grain structures are given.

MAGNESIUM ALLOYS

TEMPERATURE EFFECT. Compression and Extrusion, A. Portevin. *Metallurgist (Supp. to Engineer, Lond.)*, Nov. 1929, pp. 173-175, 2 figs.

Review of paper previously indexed from Revue de Metallurgie, Aug. 1929; relationship between behavior of given metal or alloy under high-temperature compression test and its behavior in extrusion press is discussed; author obtains autographic stress-strain diagrams of four-distinct types; interpretation of these four types in terms of single generalized pair deformation of metal under viscous flow and under work-hardening conditions respectively.

MALLEABLE IRON

ANNEALING. Effect of Cast Iron Composition on Durability of Tempering Boxes, N. G. Hirshovich. *Vestnik Metallopromishlennosti (Moscow)*, no. 7, July 1929, pp. 103-122, 9 figs.

Effect of length of life of tempering boxes made of cast iron on cost of production of malleable cast iron; results of Russian tests made at Institute of Metals and shipbuilding plant in Leningrad, to determine effect of carbon, silicon, manganese, sulphur, phosphorus, and chromium; author offers formula of best composition of cast-iron for making tempering boxes. (In Russian.)

MALLEABLE IRON CASTINGS

Malleable Iron Castings (Temperguss), B. Osann. *Maschinenbau (Berlin)*, vol. 8, nos. 23 and 24, Dec. 5 and 19, 1929, pp. 785-789 and 832-834, 20 figs.

Definition; melting, casting, and annealing in pots and furnaces, cleaning and processing, are discussed; metallurgical processes, such as admixture of manganese, silicon and sulphur, influence properties; causes of

failures in annealing; strength qualities; strength of malleable cast iron.

Molybdenum in Malleable Cast Iron, F. L. Coonan. *Heat Treating and Forging*, vol. 15, no. 12, Dec. 1929, pp. 1561-1563, 7 figs.

Possibility of using molybdenum to improve properties of malleable cast iron are discussed; from study and comparison of microphotographs, it is evident that molybdenum tends to retard decomposition of cementite into iron and temper carbon; slight increase in ductility of specimens with only small quantities of this element is probably due to refinement of ferrite; molybdenum tends to promote formation of smaller grains.

ANNEALING. Carbide Disintegration Occurring with Annealing of Malleable Cast Iron (Ueber den Karbidzerfall beim Gluehen von Temperguss), E. Schuez. *Giesserei (Duesseldorf)*, vol. 16, no. 51, Dec. 20, 1929, pp. 1185-1189, 14 figs.

Adjustment of metastable and stable equilibria at annealing temperature; influence of silicon and sulphur content on rapidity of adjusting these equilibria; annealing and quenching tests; number and size of graphite flakes at different annealing temperatures; phenomena occurring with normal quenching.

MELTING POTS

ELECTRIC. Electric Heating Units for Special Applications, R. Daniels. *Fuels and Furnaces*, vol. 7, no. 12, Dec. 1929, pp. 1933-1934, 2 figs.

Discussion of application of cast-in immersion electric heating units to low-temperatures (up to 950 degrees Fahr.) melting pots.

METALS

DEFORMATION. Plastic Deformations, Strength and Hardness of Metals (Deformaciones plasticas, tenacidad y dureza de los metals), J. Orland. *Anales de la Asociacion de Ingenieros (Madrid)*, vol. 8, no. 58, Oct. 1929, pp. 529-536, 3 figs.

Deformation produced in metals by tensile and compressive stresses are due to displacements along slipping plane of grains; effects of cold working are discussed, as regards resultant hardening, as well as effect of alternating forces.

FATIGUE: Evaluation of Fatigue Test (Beitrag zur von Dauerversuchen), K. Schaechterle. *Stahlbau (Supp. to Bautechnik, Berlin)*, vol. 2, no. 20, Oct. 4, 1929, pp. 238-239, 5 figs.

Theoretical mathematical discussion of interpretation of fatigue tests including case of alternating stresses.

GAS ABSORPTION. Hydrogen and Nitrogen Absorptivity of Pure Iron and Some of Its Alloying Elements (Ein Beitrag zur Frage der Aufnahmefaehigkeit des reinen Eisens und einiger seiner Legierungselemente fuer Wasserstoff und Stickstoff), E. Martin. *Archiv fuer das Eisenhuettenwesen (Duesseldorf)*, vol. 3, no. 6, Dec. 1929, pp. 407-416, 11 figs.

Results of investigation to determine behavior of pure iron, chromium, molybdenum, and tungsten; study of four iron-silicon alloys with 0.55, 1.68, 3.05, and 5.10 per cent silicon; gas absorption and

atomic lattice structure of metals. Bibliography.

Gas Permeability of Metals (Perméabilité des métaux aux gas), V. Lombard. *Revue de Métallurgie (Paris)*, vol. 26, no. 10, Oct. 1929, pp. 519-531, 6 figs.

Status of problem is discussed; comparison of experimental results; attempt is made to classify different systems of gas-containing metals studied.

MAGNETIC PROPERTIES. The Magnetic Behavior of Nickel and Iron Films Condensed in Vacuum Upon Various Metal Backings, J. H. Howey. *Phys. Rev.*, vol. 34, no. 11, Dec. 1, 1929, pp. 1440-1447, 5 figs.

Magnetic nature of evaporated iron and nickel films deposited on various kinds of backings was investigated to determine effect of two-dimensional strain which backing must impose upon film due to differential thermal contraction.

VACUUM MELTING. Hydrogen and Carbon Monoxide Contents of Some Metals Melted in a Vacuum (Sur la teneur en hydrogène et en oxyde de carbone de quelques métaux fondus dans le vide), A. Villachon and G. Chaudron. *Académie des Sciences—Comptes Rendus (Paris)*, vol. 189, no. 7, Aug. 12, 1929, pp. 324-326.

Metals are melted in vacuum of about 1/50 mm. at 1700 deg. or 1/500 mm. at 1000 deg. in magnesium crucibles that have previously been heated at a very high temperature; it is probable that in liquid state hydrides and carbonyl compounds having very low dissociation pressures remain dissolved in metal.

X-RAY ANALYSIS. X-Ray Metallography in 1929, G. L. Clark. *Metals and Alloys*, vol. 1, no. 5, Nov. 1929, pp. 206-221, 33 figs.

X-ray analysis of fabrication of metals and alloys; structure of electrodeposited metals, mirrors, and sputtered films; practical applications of X-rays to problems of metallurgical industry; comparison of effects of twisting and bending steel wires; effect of constitution on structure of wires drafted and annealed (basic open hearth and acid bessemer steel); twinning in annealed cold rolled steel.

METALS CLEANING

Production Cleaning, T. L. Wheeler. *Metal Cleaning and Finishing*, vol. 1, no. 7, Nov. 1929, pp. 651-656, 666.

Discussion of factors governing selection of types of cleaning equipment; application and operation of equipment.

METALS CORROSION

Oxygen's Role in the Electrolytic Corrosion-Testing of Copper and Nickel in Sodium Chloride Solutions, H. S. Rawdon and W. A. Tucker. *Metals and Alloys*, vol. 1, no. 6, Dec. 1929, pp. 277-278, 4 figs.

Accelerated electrolytic corrosion tests were made on sheet copper and nickel in sodium-chloride solution; both anode and cathode were of same metal, and applied e.m.f. was relatively very low; results are expressed as loss-of-weight-potential curves; results clearly show that accelerating effect of oxygen on corrosion by reducing potential necessary for given corrosion rate. From Bur. of Standards *Jl. of Research*, Sept.

1929, previously indexed in Engineering
Index, 1929.

METALS CUTTING

Turning with Shallow Cuts at High
Speeds, H. J. French and T. G. Digges.
U. S. Bur. of Standards—Jl. of Research,
vol. 3, no. 6, Dec. 1929, pp. 829-989, 42
figs.

Method described for testing lathe tools
under shallow cuts and fine feeds; relations
determined between cutting speed, feed,
depth of cut, and tool life for carbon and
high-speed tool steels; comparisons made of
tools of different forms and of tool life when
cutting dry and with water or lard oil; heat
treatment and chemical composition of tools
also studied, including effects of cobalt,
nickel, molybdenum, arsenic, antimony,
phosphorus, sulphur, copper, tin, aluminum,
titanium and tantalum.

Cutting and Punching of Metals (Le
cisaillement et le poinconnage des métaux),
C. Fremont. *Société d' Encouragement
pour l' Industrie Nationale—Bul. (Paris),*
vol. 128, July-Sept. 1929, pp. 563-672, 268
figs.

Historical review of art from origin to
present-day practice; development of ma-
chinery and influence of various methods
of mechanical cutting and punching on
structure of metals is shown with enlarged
photographs and curves.

METALS FATIGUE

Fatigue Characteristics of Cold Worked
Metals, H. F. Moore. *Metal Stampings,*
vol. 2, no. 11, Nov. 1929, pp. 875-876, 900.

General discussion of some of properties
of metals as shown by fatigue studies in
recent years.

METALS TESTING

The Inspection of Metals. *Aeroplane
(Lond.),* vol. 37, no. 22, Nov. 27, 1929,
(Aeronautical Engineering Supp.) pp. 1244
and 1246.

Ratio of gage-length to diameter impor-
tant; as yield point does not really exist in
certain hardened and tempered steels cast
alloys cold worked steels and alloys, proof
stress is being used to replace it; Brinell
test and machines; magnetic testing for
cracks; sulphur printing. Abstract of paper
presented before Instn. Automobile Engrs.
and Roy. Aeronautical Soc.

Vibration Strength (Schwingungsfestig-
keit), P. Ludwik. *Zeit. des Oesterreich-
ischen Ingenieur und Architekten Vereines
(Vienna),* vol. 81, nos. 41/42, Oct. 11, 1929,
pp. 403-406.

Results of bending and impact endurance
tests of polished and variously notched bars
of duralumin, copper, zinc, cast iron, steel,
steel alloys, and other metals, are discussed.

MICROSCOPES

Don't Be Afraid of the Microscope, E.
R. Schwarz. *Textile World,* vol. 76, nos.
17 and 24, Oct. 26, Dec. 14 and 21, 1921,
pp. 2430-2434 and 2445; 4364-4366 and
4405, and 4500-4503, 28 figs.

Oct. 25: Methods and equipment needed
to carry out micrometric examination of tex-
tiles. Dec. 14 and 21: Mounting textile
specimens, discussing selection of glass slides
and chemicals, and technique of mounting
specimens.

MOLYBDENUM STEEL

Molybdenum Steels Gaining Favor, E. E.
Thum. *Iron Age,* vol. 125, no. 2, Jan. 9,
1930, pp. 141-144 and 201-202, 4 figs.

Survey of present applications of molyb-
denum steel including, crankshafts, hammer
rams, die blocks, gun tubes and armor
plate.

NITRIDATION

A Double Session at Detroit, H. G. Free-
land. *Soc. Automotive Engrs. Jl.,* vol. 25,
no. 6, Dec. 1929, pp. 695-696.

Nitralloy in automotive industry is dis-
cussed; treating time being reduced; origin
and properties of nitralloy; time required
now seems to range from 1½ hour for
producing very hard phonograph needles to
30-60 hours for crankshafts and gears.
Abstract.

Observations on Iron-Nitrogen System, S.
Epstein, H. C. Croesbeck and I. J. Wymore.
U. S. Bur. of Standards—Jl. of Research,
vol. 3, no. 6, Dec. 1929, pp. 1005-1027, 26
figs.

By means of thermal analyses, micro-
scopic examinations, and X-ray analyses of
nitrided electrolytic iron specimens study
has been made of iron-nitrogen system;
from data obtained and with diagrams of
Sawyer and Fry as basis, modified iron-
nitrogen constitution diagram has been ten-
tatively drawn; upper temperature horizon-
tal noted by Sawyer was also observed, but
this has been ascribed to peritectoid instead
of eutectoid transformation; observations on
several specimens of aluminum-molybdenum
nitriding steel are also described.

OPEN HEARTH FURNACES

Insulated Open-Hearth Furnace Built.
Iron Age, vol. 124, no. 25, Dec. 26, 1929,
pp. 1737-1738, 3 figs.

Description of 25-ton open-hearth acid
furnace designed by Freyn Engineering
Co., Chicago, for operation by Falk Corp.,
Milwaukee; oil-fired unit measures 12 by
44 ft., and has average fuel consumption of
35 gallons of oil per ton of metal; excellent
operating results reported for unit encased
in ¼-inch plate to avoid infiltration; electri-
cally controlled reversing valve.

OPEN HEARTH FURNACE PRACTICE

Furnace Construction; Operation. *Iron
Age,* vol. 124, no. 23, Dec. 5, 1929, pp.
1528-1530.

Design of furnace, use of auxiliary equip-
ment, insulation, character of ports and
checker brick are considered, as discussed
at Open-Hearth Committee of American In-
stitute of Mining and Metallurgical Engi-
neers, are taken up; inspection of ladle
trunnions, cable and chains; desirable type
of soaking pits. (Concluded.)

Distinction Between Solid and Molten
Pig-Iron Charge in Open-Hearth Furnaces
(Der Unterschied zwischen festem und flues-
sigem Roheiseneinsatz im Siemens-Martin-
Ofen), E. Killing. *Stahl und Eisen (Dues-
seldorf),* vol. 49, no. 51, Dec. 19, 1929, pp.
1821-1826, and (discussion), pp. 1826-1827,
15 figs.

Results of investigation show that charg-
ing with molten pig iron gives better ton-
hour output than with solid pig; ore con-
sumption is less with solid than with liquid

charge; slag quantity is greater with molten charge, etc.

PICKLING EQUIPMENT

Longer-Lived Equipment for Metal Pickling, C. A. Crawford. *Am. Mach.*, vol. 72, no. 2, Jan. 9, 1930, pp. 58-61, 6 figs.

Discussion of corrosion-resistant alloys used in construction of conveyors, racks, and hoods of continuous and semi-continuous installations; use of forged and welded chains of acid-resistant metals with combined strength and safety of steel chains with higher and more uniform corrosion resistance; for use in its own process International Nickel Co. has developed pickling solution that can be worked hot without undue fuming and which will pickle Monel metal, pure nickel, and high-nickel white metals with good results.

POLISHING MACHINES

Metallographic Polishing, S. Epstein and J. P. Buckley. *U. S. Bur. of Standards—Jl. of Research*, vol. 3, no. 5, Nov. 1929, pp. 783-794, 9 figs. on supp. plates.

Automatic metallographic polishing machine, designed particularly for studying polishing methods, but should be useful in any metallographic laboratory; metal specimen mounted in metal ring and held in arm which moves it back and forth along radius of turning polishing disk; specimen rotated; three arms to disk so that three times as many specimens as there are disks can be polished at time; high quality polish, free from pitting and scratches obtained; proposed study of polishing methods.

PROTECTIVE COATINGS

METALLIC. Surface Protection Through Sprayed Metallic Coatings (Oberflaechenschutz durch aufgespritzte metallische Ueberzeuge), Kutscher. *Maschinenbau (Berlin)*, vol. 8, no. 16, Aug. 15, 1929, pp. 543-545, 5 figs.

Metal-spraying gun and its principal application; coatings of zinc, aluminum and aluminum-zinc; protection of iron parts against fire; spraying of cavities in cast pieces; spraying of metal powder.

METALLIC. Practical Experiences with Protection of Boiler Parts Against Combustion by the "Alumetier" Process (Praktische Erfahrungen ueber den Schutz von Kesselteilen gegen Verbrennung durch das Alumetierverfahren), C. Commentz. *Korrosion und Metallschutz (Leipzig)*, vol. 5, no. 11, Nov. 1929, pp. 248-249.

"Alumetier" process, which has been successfully used in past few years for protection of grate bars and other boiler parts from furnace gases, consists in first spraying metal parts with aluminum layer, after which they are given airtight coating; aluminum and iron are diffused at suitable temperature, thus forming tenacious heat-resisting alloy.

RAILS—HEAT TREATMENT

Heat Treatment of Rails (Tratamento Thermico dos Trilhos), J. B. da Costa Pinto. *Boletim do Instituto de Engenharia (Sao Paulo)*, vol. 11, no. 53, Oct. 1929, pp. 270-286, 12 figs.

Historical data; notes on sorbite; practice at Hagondange Shops; Neuves Maisons pro-

cess; heat treatment in United States; heat treatment in Germany; specifications of American Railway Engineers Assn.; rails for street railways.

SHEET STEEL

MAGNETIC. The Properties of Sheet Magnetic Materials, B. G. Churcher. *Elecn. (Lond.)*, vol. 103, no. 2687, Nov. 29, 1929, pp. 659-662, 5 figs.

Essential knowledge for designer; permeability of steels; effect of silicon content; hysteresis loss under alternating magnetization.

SHEET STEEL MANUFACTURE

Sheet Steel and Blackplate Manufacture, J. Mort. *Blast Furnace and Steel Plant*, vol. 17, no. 12, Dec. 1929, pp. 1816-1822, 10 figs.

Heating arrangements, roll contours, uniformity of gage and every day problems of sheet-mill man are discussed; roll temperature charts; bar furnaces; reheating furnaces; rate of heating; roll turning or grinding.

SHIPBUILDING

CORROSION RESISTING STEEL. Corrosion-Resisting Steel and its Application to Ships, H. E. Saunders. *Soc. Naval Architects and Mar. Engrs.—Advance Paper*, no. 8, for mtg. Nov. 14-15, 1929, 21 pp., 9 supp. plates. See also abstract in *Mar. Eng. and Shipg. Age*, vol. 34, no. 12, Dec. 1929, p. 682.

There are now available to designer and builder of vessels, in commercial shapes and forms, half a dozen types of corrosion-resisting alloy steel which offer means of improving design, bettering construction, increasing reliability and serviceability and raising general standard of efficiency for ships and their machinery.

SILICON STEEL MANUFACTURE

Silicon Steel for Transformer Sheets, J. H. Hruska. *Blast Furnace and Steel Plant*, vol. 17, no. 11, Nov. 1929, pp. 1652-1655, 4 figs.

Discussion of metallurgical qualities and methods in manufacture of silicon steel for transformer sheets; influence of silicon upon magnetometric properties of low-carbon steel containing silicon in various amounts; metallography of silicon transformer sheets; size of crystallites in finished transformer sheets.

STAINLESS STEEL

German Version of Birth of Stainless Steel, P. H. Schottky. *Iron Age*, vol. 124, no. 23, Dec. 5, 1929, p. 1512.

Development of stainless steel in Krupp Laboratories is traced. From *Kruppsche Monatshefte*.

STEEL

COLD WORKING. Influence of Cold Working on Magnetic Properties of Carbon Steel (Der Einfluss des Kaltreckens auf die magnetischen Eigenschaften eines Kohlenstoffstahles), W. S. Messkin. *Archiv. fuer das Eisenhuettenwesen (Duesseldorf)*, vol. 3, no. 6, Dec. 1929, pp. 417-425, 32 figs.

Influence of degree of stretching; re-

crystallization with subsequent annealing; coercive force under influence of annealing temperature; quenching tests with cold-worked steel; practical conclusions.

TEMPERATURE EFFECT. Steam Piping of the New Central Plant of C.P.D.E. and Properties of Steel at High Temperatures (La tuyauterie de vapeur de la nouvelle Centrale de la C.P.D.E., etc.), Dessus, Leconte, and J. Galibourg. *Société des Ingenieurs Civils de France—Memoires et Compte Rendu des Travaux (Paris)*, vol. 82, nos. 5/6, May/June 1929, pp. 479-538, 16 figs.

Description of high-pressure steam electric plant of Issy-les-Moulineaux of Compagnie Parisienne de Distribution d'Electricité with special reference to tests showing behavior and strength of steel of high-pressure steam pipe.

STEEL ANALYSIS

INCLUSIONS. Theoretical Considerations in the Electrolytic Determination of Non-Metallic Inclusions in Steel, C. H. Herty, G. R. Fitterer, and W. E. Marshall. *Min. and Met. Investigations—Cooperative Bul.*, no. 44, 1929, 25 pp., 4 figs.

Data given deal only with theoretical aspects of electrolytic extraction; none of aqueous extraction methods proposed and used will quantitatively recover any inclusions containing manganese oxide; inclusions extracted by electrolytic means are contaminated by atmospheric oxidation, liberation of hydrogen, metallic iron, and porous cup material.

MANGANESE DETERMINATION. Determination of Manganese in Steel by Wald's Method (Ueber die Manganbestimmung im Stahl nach Wald), J. Kassler. *Chemiker-Zeitung (Koethen)*, vol. 53, no. 74, Sept. 14, 1929, p. 719.

Modification of Volhard's method as modified by Wald is recommended for determination of manganese in chromium, nickel-chromium, and high-speed tool steels containing less than 0.2 per cent cobalt; method is described.

STEEL HEAT TREATMENT

Steel Treating Practice, R. H. Sherry. McGraw-Hill Co., N. Y., 1929, 399 pp., illus., tables, $4.00.

Book aims to outline requirements of practice rather than to discuss theory of steel treating; basic principles are presented briefly, with description of standard shop practice, and commercially important variants, in annealing, hardening, and carburizing steels; furnaces and fuels, pyrometers, control and inspection, and arrangement and equipment of treating plants are discussed in practical way; industrial aspects of subject are emphasized throughout book. Eng. Soc. Lib., N. Y.

ANNEALING. Annealing of Steel in Protective Atmosphere of Producer Gas (Glodgning av stal i skyddsgas), A. Johansson and E. von Wachenfeldt. *Jernkontorets Annaler (Stockholm)*, June 1929, pp. 141-179 and (discussion) pp. 180-189.

Results of investigation of possibility of bright annealing steel with charcoal producer gas shows that this can be done; description of apparatus used.

STEEL INGOTS

Designing Ingots for Good Surfaces. *Iron Age*, vol. 125, no. 2, Jan. 9, 1930, pp. 158-159, 7 figs.

Importance of corners in designing ingots for good surfaces is stressed in examination of action of liquid steel in mold and plastic steel in rolls; improved cross-section of ingot which solidifies without stress and rolls without overlapping at salients.

SEGREGATION. Studies of Behavior of Alloying Elements of Iron, Especially Oxygen with Segregation of Steel (Untersuchungen, ueber das Verhalten der Begleitelemente des Eisens, etc.), P. Bardenheuer and C. A. Mueller. *Stahl und Eisen (Duesseldorf)*, vol. 49, no. 50, Dec. 12, 1929, pp. 1804-1805.

Account of work carried out in laboratory of Kaiser Wilhelm Institute for Iron Research, including notes on oxygen determination; study of segregation in ingot steel molds.

Configuration of Segregation in Ingots after Solidification (Sur la configuration de la ségrégation des lingots après solidification), G. d'Huart. *Revue de Métallurgie (Paris)*, no. 10, Oct. 1929, pp. 532-537, 7 figs.

Author describes manner of crystallization of steel during cooling in ingot mold, as set out by Ingot Committee of Iron and Steel Institute and puts forward criticisms on various points; segregation and effects on it of physical and chemical factors; process of solidification; author disagrees with Benedick's statement that external segregation in inverted ingot (small end downwards) has V-shape, reverse of that found in normal ingot.

STEEL MANUFACTURE

ELECTRIC PROCESS. Production of Electric Steel for Castings, G. Batty. *West. Mach. World*, vol. 20, no. 12, Dec. 1929, pp. 451-454, 1 fig.

Two sharply differentiated methods of procedure are discussed as being applicable to production of both straight carbon and alloy steels for castings; it is affirmed that, in ordinary conditions that pertain in steel foundry, where scrap contaminated by adherent sand is used as part of charge, definite boil of bath must be secured; with some inevitable elimination of carbon; in order that finely divided nonmetallics be cleansed from bath.

STEEL RESEARCH

Steel Corporation Research Program, J. Johnston. *Iron Age*, vol. 125, no. 1, Jan. 2, 1930, pp. 61-62.

Fundamental work on behavior of iron at high temperature is supplemented by work in research laboratories and investigations at various steel plants; fundamental investigations at Kearny; research laboratories in Pittsburgh and South Chicago.

STEEL TESTING

Increase of Durability by Surface Compression (Die Steigerung der Dauerhaltbarkeit durch Oberflaechendruecken), O. Foeppl. *Maschinenbau (Berlin)*, vol. 8, no. 22, Nov. 21, 1929, pp. 752-755, 7 figs.

By compressing surface, durability of steel rods can be increased from 15 to 20 per

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MAGNETIC. Determination of Mechanical Properties of Steel From Magnetic Phenomena Observed in Drawing of Steel Bars, P. Mishin and V. Bazilevich. *Vestnik Metallopromishlennosti (Moscow)*, nos. 5 and 7, May and July 1929, pp. 19-29 and 88-102, 54 figs.

Review of research done on this problem in United States and Europe; report on original flux-meter tests by authors; effect of repeated stresses; internal stresses; results of authors' 158 tests show that method is suitable for determination of true elastic limit which was found to be function of chemical composition, mechanical and heat treatment of steel. (In Russian.)

TOOL STEEL

A Note on Production on Steels and Alloys. *Aircraft Eng. (Lond.)*, vol. 1, no. 10, Dec. 1929, pp. 342-343.

Properties of new tool steels for machining and group of cutting alloys with remarkable performances are described; hardening in gas-fired furnace; hardening in Blacksmith's hearth; quenching; secondary hardening treatment; Ajax Northrup high-frequency electric crucible process; third great advance is discovery of tungsten carbide cobalt cutting alloys.

HARDENING. Inherent Hardenability Characteristics of Tool Steel, B. F. Shepherd. *Am. Soc. Steel Treating*—TRANS., vol. 17, no. 1, Jan. 1930, pp. 90-110 and (discussion) 110, 15 figs.

Paper discusses variations in hardening properties encountered in carbon tool steels; pointing out that among steels of practically identical composition, wide variations in depth of hardness penetration are possible; it shows by use of special test, magnitude of these variations between heats from same manufacturer and between different manufacturers.

HEAT TREATMENT. Heat Treatment of Tool Steel, D. Smith. *Iron and Coal Trades Rev. (Lond.)*, vol. 119, no. 3224, Dec. 13, 1929, p. 905.

Notes on ordinary high-speed steel, with carbon 0.65, tungsten 13.08, chromium 2.97, vanadium 0.52 per cent; usual practice is to heat it in two stages before cooling; importance of method of heating; secondary hardening. Abstract of paper read before Northeast Coast Instn. of Engrs. and Shipbldrs.

INSPECTION. Notes on the Inspection of Tool Steels, C. B. Gordon-Sale. *Machy. (Lond.)*, vol. 34, no. 874, July 11, 1929, pp. 477-478.

Methods of inspecting and testing tool steel are described covering mild steels, including case-hardening steel, cast steels, including spring steel, and alloy steels, ranging from ordinary high-speed to chrome vanadium steel.

TOOL STEEL HARDENING

Procedure in Hardening Steel Tools, A. H. Hert. *Heat Treating and Forging*, vol. 15, no. 10, Oct. 1929, pp. 1291-1292, 1 fig.

Proper heating procedure in hardening tool steel is discussed; necessity for proper methods; causes of expansion; precaution in quenching; grades of hardness for various purposes; pyrometer as aid in hardening.

TOOL STEEL TESTING

Comparison of Practical and Laboratory Test of Tools of Tempered Steel (Jamforelse mellan praktiska prov och laboratorieprov pa verktyg av hardat stal), A. Lundgren. *Jernkontorets Annaler (Stockholm)*, no. 8, 1929, pp. 375-389, 7 figs.

Communication from State Testing Laboratory; veneer knives, tobacco knives, reamers, taps and threading dies have been tested in shop and laboratory; several kinds of steel and heat-treating methods have been used, and results show that machining properties of tools generally can be determined in laboratory.

TORSION TESTING

Torsion Test (Der Verdrehungsversuch), F. Meyer. *Siemens Zeit. (Berlin)*, vol. 9, no. 8, Aug. 1929, pp. 496-499, 4 figs.

Calibration equipment for statical determination of ratio between boundary deformation to boundary stresses; equipment for dynamic testing of Foeppl-Buseman system is described.

The Foettinger Torsionmeter. *Engineering, (Lond.)*, vol. 128, no. 3328, Oct. 25, 1929, pp. 542-543, 2 figs.

Indicator rotates with shaft and its dial, being brightly illuminated by proper shaded electric lamp which moves with it; there is, it is stated, no difficulty whatever in taking readings whether shaft be running fast or slow; sleeve disk and drum are all made of aluminum so that even largest sizes of torsion meter is of moderate weight; it is made in nine standard sizes.

WELDING WIRES

TESTING. Welding Wires, Their Composition and Testing Methods (Ueber Schweissdraehte, Pruefverfahren und Richtlinien fuer die Zusammensetzung. *Autogene Metallbearbeitung (Halle)*, vol. 22, nos. 17 and 18, Sept. 1 and 15, 1929, pp. 243-255, and 265-270, 31 figs.

Necessity of and specifications for physical, chemical, and microscopic testing; workshop tests, and testing of welding wires; many tables pertaining to test results, properties of welding material, etc., are given.

WROUGHT IRON

Wrought Iron—Some of Its Uses, J. S. Trinham. *Colliery Guardian (Lond.)*, vol. 139, no. 3596, Nov. 29, 1929, p. 2072; see also *Iron and Coal Trades Rev. (Lond.)*, vol. 119, no. 3220, Nov. 15, 1929, p. 754.

Wrought iron is extensively used in coalmine equipment, especially in suspension gear of mine-hoist cages, because of toughness and ability to cope with sudden shock and over-strain; examples of special uses are given.

ZIRCONIUM

Properties and Use of Zirconium Products (Ueber die Eigenschaften und Verwendung von Zirkonerzeugnissen). *Zeit. fuer die gesamte Giessereipraxis (Berlin)*, vol. 50, no. 36, Sept. 8, 1929, pp. 143-144.

Excellent refractory properties of zirconium make it of special value for use in industrial furnaces; compositions of a few zirconium products are listed.

Reviews of Recent Patents
By NELSON LITTELL, Patent Attorney
475 Fifth Ave., New York City—Member of A. S. S. T.

1,741,128, Dec. 24, 1929, Reactive Packing for Metallurgical Refractories, Charles S. Hollander, of Philadelphia, and Edward L. Helwig, of Bristol, Pennsylvania.

This patent describes a process for sealing the cracks in an electric furnace by surrounding the refractory wall with a packing which will react chemically, such as silica and alkali salt, to form a plastic mass at the prevailing temperature.

1,741,336, Dec. 31, 1929, Carburizing Material, George W. Pressell, of Philadelphia, Pennsylvania, Assignor to E. F. Houghton and Company, of Philadelphia, Pennsylvania, a corporation of Pennsylvania.

This patent relates to a relatively cheap carburizing material which has a heavy metal oxide, such as iron, chromium, nickel, cobalt and others, to bring about the increased carburizing activity. It is suggested that the oxygen is liberated from the oxide which is absorbed later, after the carburizing heat, whereby the carburizer is ready for subsequent heat. The heavy metal oxide, giving off the oxygen, converts the normal carbon monoxide in the carburizing pot into carbon dioxide, conforming to the belief that it is carbon dioxide that gives its carbon to the steel.

1,742,487, Jan. 7, 1930, Method of Manufacturing Steel, Joseph Kent Smith, of Detroit, Michigan, Assignor to Granular Iron Company, a corporation of Michigan.

This patent relates to a method of manufacturing open-hearth steel and particularly relates to the use of a protective shield to reduce the oxidizing influence of the furnace castings on the solid produced iron. The shield suggested is formed by charging on top of the usual preliminary stone charge a quantity of solid produced iron which has been agglomerated into relatively dense masses. This charge is covered with a mixture of ground limestone, ground basic slag and ground carbon. The ground limestone prevents the slag particles from coalescing too rapidly. It is found that there is little or no extra slag as the carbon passes away in a gaseous form and the limestone used takes the place of part of the limestone normally used.

1,742,986, Jan. 7, 1930, Inhibitor, Harry P. Corson, of Lakewood, Ohio, Assignor, by Mesne Assignments, to The Grasselli Chemical Company, of Cleveland, Ohio, a corporation of Delaware.

This patent relates to pickling and other acid treatments of metallic articles and is for the purpose of preventing the attack of the acid on the metal itself while removing the scale. The inhibitor used is mentioned

as cinchona bark alkaloid or a dilute mineral acid containing a small amount of quinoidine.

1,743,956, Jan. 14, 1930, Induction Furnace, Paul Emmanuel Bunet, of Versailles, France, Assignor of One-Half to Societe Acieries de Gennevilliers, of Gennevilliers, Seine, France, a corporation of France.

This patent relates to an induction furnace having an improved magnetic circuit by using magnetic iron cores externally of the primary circuit to direct the flux and to reduce the reluctance of the magnetic circuit substantially to that of the air in the part to be heated. The use of the external cores directs the flux to a particular zone of the furnace to materially increase its efficiency and the device is particularly adapted for use in tilting or other movable furnaces.

1,742,791, Jan. 7, 1930, Heat Treatment of Metals, Albert A. Somerville, of Flushing, New York, Assignor of Two-Thirds to R. T. Vanderbilt Company, Incorporated, of New York, N. Y., a corporation of New York.

This patent relates to heat treatment and particularly to oil treatment for tempering. It is to avoid the usual change in composition and deterioration of the oil baths during repeated heat treatments and it is proposed that a small amount of a condensation product of a carbonyl derivative (aldehydes or ketones) with a nitrogenous base, such as ammonia and amines, be added to the oil. The condensation product should be soluble with the oil to be intimately distributed throughout the composition and it is only necessary to use a very small amount.

1,744,213, Jan. 21, 1930, Smelting of Ferro-Alloys in Blast Furnaces, Frank W. Davis, of Milford, Delaware, Assignor to Samuel G. Allen, Trustee.

This patent relates to the smelting of ferro-alloys in blast furnaces and particularly relates to blowing the furnace with an oxygen enriched blast, admitting fuel at the tuyere zone, and by using additional oxygen admitted to the shaft to convert the potential heat in the top gas, which is CO, and transmit it as sensible heat. The addition of this oxygen enriched blast and the fuel at the tuyere zone increases the rate of smelting without increasing the blast pressure and will remove carbon at a faster rate than usual without relatively increasing the volume of gases leaving the smelting zone. It also states that it is possible to dispense with heating the blast and the stoves employed for this purpose.

1,744,374, Jan. 21, 1930, Manufacture of Stainless Iron, Walter M. Farnsworth, of Canton, Ohio, Assignor to Central Alloy Steel Corporation, of Massillon, Ohio, a corporation of New York.

This patent relates to the manufacture of stainless steel of a low carbon high chromium type and is characterized by the operation of maintaining a ferrous metal containing carbon and chromium under an oxidizing slag and at a super heat temperature, thereby oxidizing a substantial part of the carbon and chromium, the chromium oxide thus formed passing into the slag and then reducing the chromium oxide to

return a substantial amount of the chromium to the bath without removing the slag. The chromium loss is greatly reduced by subsequently reducing the chromium oxide and a good use for the large quantities of scrap normally made is found.

1,744,418, Jan. 21, 1930, Manufacture of Steel, Earle C. Smith, of Massillon, Ohio, Assignor, By Mesne Assignments, to Central Alloy Steel Corporation, of Massillon, Ohio, a corporation of New York.

This patent relates to the manufacture of steel and more particularly to the use of certain deoxidizers, such as manganese. To such a deoxidizer a frit of lead oxide and sodium fluoride is added in the proportion of one pound of frit to one ton of steel and it is found that the resulting steel has a grain size of about one-sixth of the average grain size of untreated steel. The steel is clean, having improved physical properties and a uniform grain structure.

1,745,645, Feb. 4, 1930, Iron-Carbon Alloy, Albert Sauveur, of Cambridge, Massachusetts, Assignor to the American Rolling Mill Company, of Middletown, Ohio, a corporation of Ohio.

This patent describes an alloy of carbon on a base of commercially pure iron. This is prepared by first driving off all impurities by rigorous refinement in a suitable furnace, such as an open-hearth or electric furnace. After this it is necessary to degasify the heat and then to add the carbon either under the protection of a slag and under reducing conditions. The alloy is thus produced having carbon 1 per cent, manganese not over 0.035 per cent, silicon not over 0.035 per cent, phosphorus not over 0.01 per cent and sulphur not over 0.03 per cent. Such a steel will take a more drastic quenching without cracking and will be softer, more ductile, possess greater resistance to shock, to fatigue and to corrosion, and will also have greater electrical conductivity and greater weldability.

1,742,286, Electrical Furnace, Harold N. Shaw, Milwaukee, Wisconsin, Assignor By Mesne Assignments, to Globar Corporation, Niagara Falls, New York.

In this electrical furnace the heat-producing body portion of a resistance heating element is located within the furnace chamber and the terminal portion of increased conductivity is disposed within an aperture in a wall of the chamber. A refractory tube within the aperture surrounds the terminal portion of the heating element. The heating element is connected to a supply circuit by means of a terminal piece of heat-resistant conductive metal having its inner end recessed and disposed within the refractory tube. The terminal piece is maintained in forcible engagement with the end of the terminal portion in the recess of the terminal piece. A cooling jacket is carried by the wall of the furnace and has an inner tubular wall surrounding and closely adjacent to the refractory tube.

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